Residual stress effects on the fracture toughness behaviour of a narrow-gap austenitic stainless steel pipe weld

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Abstract

Residual stress effects on the fracture toughness behaviour of a narrow-gap austenitic stainless steel pipe weld

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Automated narrow-gap girth-butt welds are replacing conventional welding methods to join sections of austenitic stainless steel pipe in the primary circuit of Pressurised Water Reactors, to reduce manufacturing costs and improve quality. To ensure the safe operation of these systems, reliable structural integrity assessments have to be undertaken, requiring the mechanical properties of welded joints to be characterised alongside the weld residual stress magnitude and distribution.

This research project characterised, for the first time, the weld residual stress field and the tensile and ductile fracture toughness properties of a 33 mm thick narrow-gap 304L stainless steel pipe weld.

The residual stress was characterised using two complementary approaches: deep hole drilling and neutron diffraction. A novel neutron diffraction scanning technique was developed to characterise the residual stress field, without cutting an access window into the component, leaving the original weld residual stress field undisturbed. A modified deep hole drilling technique was developed to characterise the residual stress retained in fracture mechanics specimens extracted from the pipe weld in two orientations. The modified technique was shown to measure the original weld residual stress field more accurately than through conventional deep hole drilling. Residual stresses, exceeding 50% of the weld material proof strength, were retained in axially-orientated fracture mechanics specimens.

Tensile tests showed that the weld was approximately 60% overmatched. It was demonstrated that neither retained residual stress, nor specimen orientation, had a discernible effect on the measured fracture toughness of the weld material. In less ductile materials, however, the level of retained residual stress may unduly influence the measurement of fracture toughness. At initiation, the fracture toughness properties of both the parent and weld materials were far in excess of the measuring capacity of the largest fracture mechanics specimens that could be machined from the weld.

The influence of residual stress and fracture toughness on the performance of narrow-gap welded pipework was investigated. Full elastic-plastic finite element analyses were used to model the pipe weld, containing a postulated defect under combined primary and secondary loading. The results, applied within the framework of an R6 structural integrity assessment, compared different plasticity interaction parameters on the prediction of failure load; the conventional $\rho$-parameter approach was compared with the recently developed, more advanced, $g$-parameter. It was shown that the $g$-parameter significantly reduced the conservatism of the $\rho$-parameter approach. However, for this pipe weld, plastic collapse was predicted to precede failure by ductile initiation, suggesting that a plastic collapse solution may be an appropriate failure criterion to use in structural integrity assessments of similar component and defect combinations.
Declaration

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1 Introduction

The Pressurised Water Reactor (PWR) is a type of light water reactor that has been in use since the mid-20th Century, as a means of generating electrical energy for use in civil electrical supplies and as a propulsion system for nuclear submarines. The first civil PWR, a demonstration reactor at Shippingport, Pennsylvania, became operational in 1957; the first fully commercial PWR, Yankee Rowe, Massachusetts, became operational in 1960 (World Nuclear Association 2011).

In a PWR, water is used as both a neutron moderator and coolant. Water is heated in the core inside the Reactor Pressure Vessel (RPV), through thermal contact with the nuclear fuel. Figure 1-1 shows the basic layout of a typical PWR.

The water, which becomes superheated, is pumped around the primary circuit at temperatures of approximately 300°C and a pressure of approximately 150 bar (~15 MPa). The primary circuit consists of the RPV and a network of pipes, which pass through a steam generator, allowing heat exchange to take place from the primary circuit water to the secondary circuit. Separate primary and secondary circuits confine the radioactive coolant to the primary circuit. Feedwater delivered through the secondary circuit is heated in the steam generator to produce saturated
steam; the steam is dried and flows along the secondary circuit pipes to drive a
turbine. In civil nuclear power plants, the turbine drives a motor generator,
producing electricity (World Nuclear Association 2011).

The pressure vessel and piping are designed to withstand the high temperature and
pressure of the water they contain. The structural integrity of these components is
essential for the safe operation of the nuclear power plant. For example, a structural
failure in the primary circuit, leading to a Loss of Coolant Accident (LOCA), can have
serious consequences (World Nuclear Association 2012), including:

- The uncontrolled release of superheated water,
- The release of radioactive matter into the environment,
- A loss of cooling capacity in the reactor core, which, if not restored,
  can lead to severe damage of the core and nuclear fuel,
- A loss of mechanical power. In naval power plant, this can result in a
  loss of propulsion and electrical supply to the life-support systems;
  auxiliary systems have limited capacity to maintain electrical supplies
  and provide propulsion.

The primary circuit pipework in a PWR is usually made of austenitic stainless steel,
chosen for a combination of properties: its general corrosion resistance, adequate
strength, ease of fabrication and high fracture toughness. The level of carbon in the
material is carefully controlled; high levels of carbon can lead to the formation of
chromium-rich carbides, which precipitate along grain boundaries. This leads to the
depletion of chromium adjacent to the grain boundaries to levels below
approximately 12%, where the material no longer exhibits corrosion protection
passivation. These depleted areas are preferentially attacked in a corrosive
environment, a phenomenon known as sensitisation (Roberts 1981). Stainless steel
grades 304L and 316L are commonly used in primary circuit pipework; the “L”
designating the low carbon content of the material (<0.03%), reducing
susceptibility to sensitisation. Additionally, grade 316L contains molybdenum,
which improves resistance to pitting and crevice corrosion in the primary circuit
water environment (Higgins 1983).
Sections of pipe, valves and fittings are joined together using circumferential girth-butt welds, through techniques such as Manual Metal Arc (MMA) or Tungsten Inert Gas (TIG) welding (Ould et al. 2009). Pipe welds are of particular concern to structural engineers; over many years of operating welded structures, most structural failures have taken place in welds or very close to them (Dzioba & Neimitz 2007). The welding process causes a large variation in the local microstructural and mechanical properties of the parent and weld material. The differential rates of contraction upon cooling create a residual stress field across the weldment (Rakin et al. 2008), that is, a stress field which remains in a component when no external load is applied. Fit-up stresses, where the restraint of a larger structure is overcome to join two components together, can also contribute to the residual stress field. Weld residual stresses can have a magnitude as high as the yield stress of the material (Lidbury 1984).

Residual stresses can contribute to the initiation of defects in PWR plant through mechanisms such as Environmentally Assisted Cracking (EAC). For example, a repair weld in the VC Summer nuclear power plant in the USA created large tensile residual stresses in a primary circuit pipe. The residual stress, combined with the environmental conditions, led to Primary Water Stress Corrosion Cracking (PWSCC), a form of EAC, which caused an axial crack to grow through the entire thickness of the primary circuit pipe wall, shown in Figure 1-2 (Clary et al. 2001).

![Figure 1-2 - Stress corrosion crack which grew through the thickness of a primary pipe weld in the VC Summer power plant. OD = Outer Diameter and ID = Inner Diameter (Clary et al. 2001).](image-url)
It is therefore desirable to remove as much weld residual stress from a structure as possible. One approach, known as Post-Weld Heat Treatment (PWHT), can relax residual stresses to levels of approximately 10-25% of the material yield stress (Lidbury 1984). However, in stainless steels, the PWHT temperature, typically 600 - 650°C, causes chromium-rich carbides to precipitate along grain boundaries, leading to sensitisation of the steel. PWHT has thus proved unsuitable for the relief of residual stress in austenitic stainless steel PWR pipe welds. Plant operators have therefore had to manage stainless steel welded structures containing high residual stress fields which combine with stresses due to service loading (Roberts 1981).

A variety of approaches have been developed over the years to measure the residual stress field in welded components. Diffraction techniques, such as Neutron Diffraction (ND) and X-Ray Diffraction (XRD), are capable of measuring residual stresses non-destructively, but can fail to measure residual stress levels deep within the bodies of large components. Such techniques often require the use of large, dedicated facilities, making measurements on in-service components impractical (Hutchings et al. 2005). Mechanical strain relief techniques, such as Deep Hole Drilling (DHD), have been developed to measure residual stress in large components, but are generally destructive or semi-destructive in their approach (Kingston & D. J. Smith 2005). Numerical modelling of the weld process, to predict residual stress fields, has been developed over recent years, but relies on validation against experimentally obtained data (Liu et al. 2011). As a result, a complementary combination of techniques is often used to characterise the residual stress fields in new component designs.

Structural integrity assessment procedures have been developed to account for the effects of residual stress combined with service stresses on defect behaviour, so that appropriate safety margins and justifications can be made for the continued operation of structures containing welds. Procedures such as R6 (British Energy Generation Limited 2000), API-RP579 (American Petroleum Institute 2000), BS7910 (British Standards Institution 2005b) and RSE-M (Faidy 2000) have been developed for such purposes.
Conducting a structural integrity assessment of a component requires information relating to the following:

a) the geometry of the component and the defect,
b) the stress state in the component, including the magnitude and distribution of residual stress as well as stresses due to service loading,
c) material properties including yield, flow and fracture properties.

Dong and Brust (2000) explain that the difficulty of reliably characterising weld residual stress fields has led to the development of assessment procedures which assume upper-bound residual stress solutions, such as API-RP579 (American Petroleum Institute 2000) and BS7910 (British Standards Institution 2005b). Such a conservative approach has often led to the effects of residual stress on the structural integrity of a component being overestimated. From the point of view of a utility or power plant operator, the reactor downtime that can be caused as a result of increased surveillance, repair and replacement of otherwise sound structures has caused unacceptable loss of plant availability (Roberts 1981). Work has been undertaken to address the conservatisms of current assessment procedures. For example, James et al. (2009) have developed a new parameter for use in the R6 procedure, to account for the interaction of residual stress with service stresses in ductile materials, where the existing R6 approach has been shown to be over-conservative.

Measuring the fracture toughness of a weld metal for structural integrity assessment involves the extraction of test pieces from a welded component. It is generally assumed that the residual stress field is relaxed on extraction of test specimens. However, recent work by Altenkirch et al. (2009) has shown that a significant level of residual stress can remain in test-pieces extracted from thin welded plates. Residual stress retained in a test specimen may influence the measured fracture toughness by contributing to crack opening forces. In the presence of tensile residual stress this may result in a measured fracture toughness lower than would be obtained from a similar, stress-relieved specimen. There is little knowledge of how the orientation of a test specimen with respect to a weldment influences the level of residual stress retained in the test specimen. Consequently, the retained residual stress can influence a structural integrity
assessment of a welded component, where the effect of residual stress may be accounted for twice: first, through the measured value of fracture toughness, and second, through the crack driving force term.

Conservatisms in existing structural integrity assessment procedures have become more apparent with improvements in manufacturing methods and welding technology. New welding techniques have been developed to mitigate the effects of weld residual stress on component reliability. Automated welding methods have improved the quality and consistency of welded joints by virtually eliminating welding-induced defects, such as weld strikes, which often acted as potent crack initiators. Weld filler alloys have been developed to minimise the contraction effect upon cooling of a weld, reducing the strength of the weld residual stress field (Shirzadi et al. 2009).

Narrow-gap welding technology has begun to replace the conventional welding methods used in nuclear plant. Often an automated process, narrow-gap welding requires less filler metal, which in turn requires less heat input, reducing the size of the Heat Affected Zone (HAZ), reducing the level of sensitisation in the HAZ (Engelhard et al. 2000). A lower required heat input also allows for an increased welding speed and less thermal contraction on cooling, reducing the size and extent of the weld residual stress field (Chapman et al. 1997). The narrow-gap welding process reduces welding time and volume of weld metal, reducing cost and increasing the quality of the welded joint. Narrow-gap weld technology is currently being used in nuclear new-build projects. However, relatively little information on the residual stress field and mechanical properties of these welds is available, which is required for accurate structural integrity assessments to be made.
The aim of this work is to develop new information and understanding of the key factors that influence the structural integrity assessment of austenitic stainless steel narrow-gap pipe welds in primary nuclear pipework. These factors include the weld residual stress field and the aspects that influence the characterisation of the material yield, flow and fracture toughness properties. Therefore, the aims of this research project are as follows:

- To characterise the weld residual stress field in an austenitic stainless steel narrow-gap pipe weld,
- To characterise the tensile and fracture toughness properties of the parent and weld materials,
- To assess the factors that influence the measured fracture toughness, including the retained residual stress in a fracture mechanics specimen, the orientation of the test specimen with respect to the weld line and the size of the test specimen,
- To develop new understanding of the factors that influence the structural integrity assessment of postulated defects in narrow-gap welded primary pipework, including approaches to account for the influence of weld residual stress on the crack driving force.
2 Literature review

The literature review first defines residual stress and its origins, focusing on weld residual stress. A range of residual stress measurement techniques is described, focusing in detail on diffraction and hole-drilling methods. This is followed by a description of the yield behaviour of materials and an introduction to the key concepts behind fracture mechanics. The principles which underlie structural integrity assessment procedures are explained, focusing on the R6 defect assessment procedure. This is followed by an account of recent test programmes investigating the properties of primary nuclear pipe welds.

2.1 Definition and origins of residual stress

For the purpose of structural integrity assessment, the stresses acting on structures are classified by their origin. Primary stresses arise from the mechanical loading of a structure, generally the loads that the structure was designed to bear. Secondary stresses can originate from a range of sources, such as residual and thermal stresses. Dong (2008) defines residual stress as the stress which remains in a body when no external mechanical load is applied. Consequently, residual stresses are sometimes referred to as “locked-in” stresses (Palanichamy et al. 2009).

Residual stress can be created in a number of ways. Fabrication processes, such as forging and rolling, can introduce residual stress into a component. Joining processes, such as welding, create residual stress. “Fit-up” residual stresses occur when several components are joined together, forming part of a larger structure. Regardless of origin, at a more fundamental level, residual stress arises from strain misfits between different regions of a material (Withers & Bhadeshia 2001b). The scale of such misfits affects the wavelength, $\lambda$, over which the residual stress self-equilibriates. Residual stresses are therefore classified by the length-scale over which they operate.

Type I stresses – sometimes referred to as macrostresses – have a wavelength ranging from the order of micrometres to metres in scale, spanning microstructural to structural distances. Engineering analyses often focus on Type I residual stresses, apparent at the continuum level, neglecting the microstructural effects of residual
stress (Bouchard & Withers 2004). Type II residual stresses vary over the scale of grains in the material, where \( \lambda \) ranges from three to ten times the grain size. Type II stresses originate from the different thermal and elastic properties between different orientations of grains in a polycrystalline material. Type III residual stresses are effective over the atomic scale, where \( \lambda \) is less than or equal to the grain size of the material. Such stresses are formed at grain interfaces or at dislocations. Bouchard and Withers (2004) describe how misfits at one scale can create residual stress effects over longer length scales. For example, a large volume fraction of Type II misfits can be significant enough in size when combined to be classified as a Type I residual stress. It is therefore important to understand how material microstructure influences the total residual stress in a component. Figure 2-1 shows schematically how the three types of stress may exist in a metal.

![Figure 2-1 - Typical residual stress length scales (Withers & Bhadeshia 2001b). \( \sigma_I \) is the type I residual stress, acting over the continuum level. \( \sigma_{II} \) is the type II residual stress, operating over the scale of grains. \( \sigma_{III} \) is the type III residual stress, acting over atomic length scales.](image)

Residual stress is of interest to materials scientists and engineers because it can affect the performance and longevity of a component in service. Residual stresses can enhance material performance: for example, compressive residual stresses can be created in the outer surfaces of a component. A defect in the outer surface is
subjected to compressive residual stress, which resists crack opening forces due to service loading (Withers & Bhadeshia 2001b). Examples of introducing compressive residual stress to improve material performance are the peening of metals and the heat treatments used to manufacture toughened glass.

Residual stress can also be detrimental to the performance of a component. Tensile residual stresses in particular can combine with the primary stresses caused by service loading, to reduce crack initiation life and increase the growth rate of defects present in a component, which can lead to premature failure (Bouchard & Withers 2004). Therefore the characterisation of the magnitude and profile of a residual stress field, and the understanding of its effects on crack initiation and growth, is necessary for accurate and reliable structural integrity assessments to be made.

### 2.2 Weld residual stress

Arc welding of metals involves localised, intense heat being used to deposit molten filler metal between two components, causing the weld material to fuse with the parent material on either side of the joint. The material surrounding the weldment experiences elastic, plastic and creep deformation (Edwards et al. 2008). Large microstructural gradients can exist across a weldment, which can lead to a large variation in material properties across the joint. When a weld cools, a residual stress field is created in the zone around the weldment. This is caused by the restraint of the structure, differential rates of contraction of the parent and weld materials, and misfits between the grains of the different materials (Rakin et al. 2008).

The size and nature of the residual stress field depends on a number of factors, such as the geometry of the joint and the welding parameters used. A range of weld geometries exist; Figure 2-2 shows five common weld joint designs and Figure 2-3 shows six common preparations for butt joints, including bevelled grooves, V-grooves and J-grooves.
The welding parameters affect the heating and cooling of the joined materials. The heating and cooling cycles to which the metals are exposed vary according to the number of passes, the path of each weld pass, the amount and rate at which filler metal is deposited, the heat input and the chemical composition of the filler metal.
These variables also affect the level of microstructural variation in the welded joint. Therefore assessing the structural performance of a welded joint can be a very complicated process (Mochizuki & Toyoda 2007). The circumferential butt weld, a joint type commonly used in nuclear power plant piping systems, often requires several weld passes due to the relatively large wall thickness of the pipes. For example, Yang et al. (2011) describe a typical butt weld in the primary circuit of a Korean PWR, having a wall thickness of 76 mm, requiring 41 weld passes to fabricate the joint.

A development of the butt-welding technique is the Narrow Gap (NG) weldment. The width of the joint is narrower than for conventional butt welds, as shown in Figure 2-4.

![Figure 2-4 - Cross-section of a typical pipe girth-weld (left) and a narrow-gap girth weld (right) (Bouchard 2007).](image)

The NG weld process is often automated, due to the restricted access of the narrow-gap geometry. A track is fitted to the outer surface of the plate or pipe, running around its circumference. The weld torch and filler metal-feed run around the track, following an orbital path. The NG weld requires less filler metal and a lower heat input, reducing the size of the HAZ and causing less thermal contraction on cooling than a conventional weld (Engelhard et al. 2000). This in turn reduces the magnitude and extent of the weld residual stress field when compared to a conventional weld. Examples of NG weld residual stress fields are shown in Section 2.8.

Engelhard et al. (2000) explain that a reduction in axial shrinkage as a result of the NG process can introduce compressive residual stresses at the weld root on the
inner diameter of a pipe. Combined with the reduced sensitisation of the HAZ, NG welds are generally less susceptible to environmentally assisted cracking, such as Stress Corrosion Cracking (SCC). Defects located in the weld region on the inner surface of the pipe experience compressive stresses at the weld root, encouraging crack closure.

Automated welding also reduces the risk of weld metal defects being introduced to the weldment, as may be the case with conventional manual welding techniques. From a plant operator’s point of view, the increased speed at which NG welding can be completed reduces the time, and hence cost, required to construct and repair piping systems (Yang et al. 2011).

Weld residual stresses can reach levels as high as the yield stress of the parent material. These secondary stresses can combine with primary stresses, increasing the susceptibility of a component to fatigue, SCC and fracture (Lidbury 1984). Figure 2-5 shows the longitudinal and transverse residual stress profile measured across a 304 stainless steel butt-welded plate. The longitudinal stresses are tensile around the weldment, reaching a peak value of approximately 350 MPa on the weld centreline. Further away from the weld centreline, the residual stress becomes compressive, balancing the tensile residual stress.
Sattari-Far and Farahani (2009) developed a Finite Element (FE) model of a 304 stainless steel girth-butt pipe weld, Figure 2-6, which shows a residual stress profile typical of such a component. A region of tensile residual stress lies close to the weld centreline; for this particular pipe weld, up to 20 mm from the weld centreline. Beyond this point, the residual stress becomes compressive, up to 65 mm, beyond which the majority of the residual stress diminishes.
2.3 Measuring residual stress

In order to perform an accurate safety assessment, the residual stress field in a welded structure must be quantified. However, measuring such stresses in a component is not a straightforward task. A range of residual stress measurement techniques have been developed. This section briefly details several approaches, followed by a more detailed description of the Neutron Diffraction (ND) and Deep Hole Drilling (DHD) techniques.

2.3.1 Centre-hole drilling

Centre-hole drilling is a mechanical strain-relief technique, capable of measuring residual stresses at the surface of a component. A hole is drilled into the surface of a component, causing elastic strain in the surrounding material to relax. The strain released can be measured using a rosette of strain gauges, shown in Figure 2-7. The strain relaxation is related to the local residual stress field through elastic equations. However, the technique has limitations; if the depth of the hole exceeds its diameter or residual stresses are greater than 50% of yield – causing localised yielding around the hole – measurements can be inaccurate (Withers & Bhadeshia 2001a).

![Rosette of strain gauges used to measure the relaxation of material around a drilled hole](Withers & Bhadeshia 2001a)

2.3.2 Ultrasonic techniques

Palanichamy et al. (2009) explain how the principle of acoustoelasticity – the stress dependence of ultrasonic wave velocity in a material – can be used to quantify residual stresses on the surface and in the subsurface of a component. Ultrasonic
waves of frequency in the 2-3 MHz range are introduced to a material through transducers. An acousto-elastic constant for unstrained material is obtained. This is compared to the speed of the wave in strained material, from which the residual stress can be deduced. However, the technique is limited to residual stress measurements close to the surface of the material, as the technique does not accurately measure residual stress in the presence of multiaxial stress fields (Withers & Bhadeshia 2001a).

### 2.3.3 Transmission Electron Microscopy (TEM)

Roy et al. (2007a) characterised internal stresses across weldments using Transmission Electron Microscopy (TEM) to measure the dislocation density in metallic materials. Higher dislocation concentrations were associated with higher residual stresses. For example, high dislocation concentrations were found in the HAZ of a weldment compared to the parent material, where high residual stresses are typically found. However, this residual stress measurement technique is of limited use as it does not quantitatively measure residual stress, nor indicate the direction in which it is acting.

### 2.3.4 Positron Annihilation Spectroscopy (PAS)

Roy et al. (2007b) describe PAS as a non-destructive method, in which a specimen is irradiated with gamma rays, inducing radioactive decay within the sample. One of the decay products is the positron, which, when electrons are fired at the sample, is annihilated at sites where neutrons are present, forming two photons. High photon counts indicate the presence of a void; low counts reveal a defect-free area of the atomic lattice. The density of voids reflects the stress levels in a particular region of the sample. However, Roy et al. (2007b) point out that the technique is not capable of differentiating between tensile and compressive residual stress and can only estimate the stress as an average value throughout the sample, with no indication as to which direction the stress is acting.
2.3.5 Contour method

The contour method is a destructive technique which takes place over three stages, outlined in Figure 2-8. A specimen containing residual stress is cut, using a method which does not introduce new stresses into the material, such as Electro-Discharge Machining (EDM). The cut faces displace locally, according to the level and distribution of residual stress acting normal to the free surfaces. The cut faces are then contour-mapped using a co-ordinate measuring machine. The measurements are applied to a Finite Element (FE) model, from which forces are applied to displace the free surface back to its original profile prior to cutting (Turski & Edwards 2009). The numerical model is then used to calculate the stresses required to make the contour flat: the calculated stresses will be the same in magnitude, but opposite in direction, to the stresses present in the component before it was cut.

The technique can be performed in many laboratories, as the equipment required is straightforward to obtain and use. It is a relatively inexpensive technique and can produce results in a relatively short space of time. It is capable of measuring a full cross-sectional profile of residual stress in a specimen. However, the contour method is destructive, preventing application of the technique to in-service components.

![Figure 2-8](image_url)

Figure 2-8 – An illustration of the contour method. The specimen is cut and the free surface is mapped. An FE simulation is used to determine the residual stress originally present by calculating the stresses required to return the contour to its original planar shape (Withers et al. 2008).
2.3.6 Diffraction techniques

The technique of diffracting electromagnetic radiation through crystalline structures was pioneered by W.H. Bragg and W. L. Bragg (1913). Originally using X-rays, diffraction was used to investigate the structure of crystalline powders and solids. A monochromatic, parallel and coherent beam of electromagnetic radiation, with wavelength $\lambda$, is incident upon two parallel planes of atoms, at an angle, $\theta$. This is illustrated in Figure 2-9, where the planes of atoms are labelled $x-x'$, $y-y'$ and $z-z'$. For clarity, only two incident beams, labelled 1 and 2 respectively, are shown. The two rays are scattered by atoms at positions $O$ and $B$. Ray 2 travels further than ray 1 upon diffraction. The difference in path length of the two beams is marked by the points $A$, $B$ and $C$. Constructive interference of beams 1 and 2 occurs if the path $ABC$ equals an integer number of wavelengths, $n$, i.e., $n\lambda = AB + BC$.

![Diagram of diffraction](image)

Figure 2-9 - Diffraction of electromagnetic radiation on the atomic planes $x-x'$ and $y-y'$.

The difference in path length can be expressed in terms of the interatomic spacing, $d_{hkl}$, Equation 2-1.

$$n\lambda = d_{hkl}\sin\theta + d_{hkl}\sin\theta$$  \hspace{1cm} 2-1
Equation 2-1 simplifies to Equation 2-2, Bragg’s law (W. H. Bragg & W. L. Bragg 1913).

\[ n\lambda = 2d_{hkl}\sin\theta \]  

2-2

The constructive interference of the diffracted beams is plotted on a graph of intensity versus diffraction angle, \( \theta \), as a series of peaks. Each peak relates to a crystallographic plane in the material.

For diffraction experiments, the diffraction angle is fixed and the wavelength of the incident beam is known, enabling \( d_{hkl} \) to be measured. The lattice-plane spacings for a component with residual stress present, \( d \), are measured. A second set of measurements is undertaken on a similar material without residual stress present, to obtain the strain-free lattice parameter, \( d_0 \). The strain due to the residual stress field, the residual strain, \( \varepsilon_{res} \), can be calculated by comparing \( d \) and \( d_0 \), or measuring the peak shift, \( \Delta\lambda \), shown in Equation 2-3 (Hutchings et al. 2005).

\[ \varepsilon_{res} = \frac{d - d_0}{d_0} = \frac{\Delta\lambda}{\lambda} - \cot \theta \cdot \Delta \theta \]  

2-3

The principal strain directions in a component are deduced from symmetry arguments, requiring three strain measurements to be made, one for each principal direction. It is then possible to calculate the principal residual stresses at each sampling location (Withers & Bhadeshia 2001a).

Various sources of electromagnetic radiation can be used in diffraction experiments. The choice of source is based on a compromise between spatial resolution and penetration depth, as well as practical considerations. Electrons can be used as a radiation source, achieving high spatial resolution, but are restricted to metallic diffraction samples of under 100 nm thickness. Thicker samples attenuate the electron beam to undetectable levels; electrons are scattered readily in metals (Hutchings et al. 2005). X-ray Diffraction (XRD) is capable of greater penetration depths, typically of the order of tens of micrometres, with good spatial resolution. It is possible to perform XRD in the laboratory, without the use of dedicated central facilities. Hard X-rays (high energy X-rays) from synchrotron sources are capable of
greater penetration depths, with fast data acquisition times, but are available only at dedicated facilities. Neutrons are the favoured electromagnetic radiation source for measuring residual stress in engineering components, as they are capable of penetration depths of the order of centimetres (Withers & Bhadeshia 2001a). However, spatial resolution is not as high as XRD, and neutron sources can only be found at dedicated central facilities.

2.3.6.1 The $\sin^2 \psi$ method

The shallow penetration depths of X-rays in metals make the XRD technique suitable for measuring stresses in the surfaces of metallic materials. At the surface of a component, the stresses perpendicular to the surface are zero. This makes it possible to measure residual stresses close to the surface, without the need for a strain-free reference sample.

![Diffraction planes at an angle $\psi \phi$ in a specimen in a biaxial stress state](image)

Figure 2-10 shows the surface of a material with a biaxial stress state present. $\sigma_{11}$, $\sigma_{22}$ and $\sigma_{33}$ are the principal stress directions. For a stress acting in the plane of the surface at an arbitrary angle, $\phi$, X-rays incident at an angle, $\psi$, will diffract with the atomic planes perpendicular to the $\phi \psi$-plane.
In terms of the principal stress directions, Dölle (1979) and Noyan (1985), show that a general expression can be used to relate a triaxial stress state and the lattice plane spacing in a general direction \(d_{\phi\psi}\), Equation 2-4.

\[
d_{\phi\psi} = d_0 \left[ 1 + \frac{1 + \nu}{E} (\sigma_{11}\cos^2\phi + \sigma_{12}\sin2\phi + \sigma_{22}\sin^2\phi - \sigma_{33})\sin^2\psi \right. \\
+ \frac{1 + \nu}{E} \sigma_{33} - \frac{\nu}{E} (\sigma_{11} + \sigma_{22} + \sigma_{33}) \\
+ \frac{1 + \nu}{E} (\sigma_{13}\cos\phi + \sigma_{23}\sin\phi)\sin2\psi \right] 
\tag{2-4}
\]

However, for a biaxial stress state, where the stress components \(\sigma_{13}, \sigma_{23}\) and \(\sigma_{33}\) are zero, Equation 2-4 can be simplified to Equation 2-5. An assumption is made that, close to a free surface, the stress-free spacing, \(d_0\), will be the same as the out-of-plane lattice spacing, \(d_{33}\), to within an error of less than 2% (Shackleton 2001).

\[
d_{\phi\psi} = d_{33} + \sigma_\phi \left(\frac{1 + \nu}{E}\right)\sin^2\psi 
\tag{2-5}
\]

Measuring \(d_{\phi\psi}\) over a range of \(\psi\) angles, the linear relationship between \(d_{\phi\psi}\) and \(\sin^2\psi\) can be plotted. From Equation 2-5, the gradient of such a plot will yield the in-plane stress, \(\sigma_\phi\). Extrapolating the plot to \(\sin^2\psi = 0\) and \(\sin^2\psi = 1\) yields the lattice spacings in the principal directions, from which the principal strains and stresses can be calculated.

### 2.3.7 Deep Hole Drilling

The Deep Hole Drilling (DHD) technique is a semi-destructive mechanical strain-relief method for measuring residual stress in engineering components. It has been continuously developed since the early work of Zhdanov and Gonchar (1978) and Beaney (1978), over the last four decades (D. J. Smith et al. 2000). An advantage of DHD over diffraction techniques is its ability to measure residual stresses at much greater depths in thick section components: Kingston and Smith (2005) claim that DHD can measure residual stress up to 450 mm away from a free surface. The apparatus required to perform DHD is relatively portable and inexpensive; DHD can be performed on-site, without requiring the use of dedicated facilities, as would be
the case with neutron diffraction. As a result, DHD can be applied to in-service components during routine shutdowns.

The DHD process is outlined by Ficquet et al. (2005), and illustrated in Figure 2-11. Reference bushes are attached to the outer surfaces of a component containing residual stress. The bushes mark the start and end points for a reference hole, which is gundrilled through the component (Figure 2-11a). The typical diameter of a reference hole is 1.5 mm. It is assumed that the reference hole is sufficiently small, and the body of the component sufficiently large, that the drilling of the reference hole does not cause relaxation or redistribution of the residual stress field.

The bore of the reference hole is measured along its entire length, using an air probe (Figure 2-11b). A jet of air is sent through the probe, the pressure of which centres the probe in the reference hole. The relationship between the air pressure and the hole diameter is established by recording the air pressure required to centre the probe in calibration discs containing holes of known diameter. The method is sensitive to the ambient conditions; the air pressure gauge is often recalibrated every few hours to maintain accuracy.

A core of material, co-axial with the reference hole, is trepanned by EDM, shown in Figure 2-11c. The residual stress contained in the material between the reference hole and the trepan relaxes, causing the diameter of the reference hole to change, illustrated in Figure 2-12.
The DHD technique is based on two assumptions: that the stress relaxation is elastic, and that all of the residual stress in the trepanned zone has fully relaxed (Mahmoudi et al. 2009). Finally, the bore of the reference hole is re-measured using...
the air probe. The change in diameter is related to the relaxed residual strains, from which the original residual stress is calculated (Ficquet et al. 2005).

The DHD technique has several limitations. It is not possible to accurately measure the component of residual stress parallel to the reference hole, because only the principal components of strain, and related shear strains, in the plane perpendicular to the reference hole, can be measured, illustrated in Figure 2-13.

![Figure 2-13 - Illustration of the stress plane measured by DHD, indicated by the grey zone with red arrows.](image)

It is possible to estimate the component of residual stress parallel to the reference hole by measuring the shrinkage of the trepanned core surrounding the reference hole. It has also been shown that the diameter of the trepanned core (and thus gauge volume) can affect the calculated residual stress. Mahmoudi et al. (2010) investigated the effect of different gauge volumes on measured residual stresses using Finite Element Analysis (FEA) to simulate the DHD technique. Using different core sizes of 3-, 10-, 20- and 50-times the diameter of the reference hole yielded different measures of residual stress in an aluminium alloy cylinder under equibiaxial tension. The core three times the diameter of the reference hole gave results in closest agreement with those predicted by the numerical model.
2.3.8 Summary

A range of residual stress measurement techniques have been developed over the last century. The choice of technique depends on the context in which residual stress measurement is required. For example, a destructive technique such as the contour method, may be appropriate for a laboratory sample or a mock-up; however it is unsuited to measuring the residual stress of an in-service component. Non-destructive techniques, such as diffraction, can be used to measure undisturbed stress fields with high resolution. Semi-destructive techniques, such as DHD, balance resolution with the flexibility to measure residual stress in components, without the need to use dedicated facilities. A review of residual stress measured in austenitic stainless steel pipe welds is given in Section 2.8.

2.4 Plasticity in metals

Under an applied tensile stress, materials such as austenitic stainless steels do not show a definite elastic limit or yield point (Higgins 1983). To characterise the yielding behaviour of such materials, the stress required to produce a definite measure of permanent extension, the proof stress, is specified. Typically, the stress required to produce 0.2% plastic strain is used, known as the 0.2% proof strength (Higgins 1983). Deformation in austenitic stainless steels, even at relatively low stresses, is often made up of a combination of elastic and plastic strains; this section discusses plasticity in the context of these metals.

2.4.1 Plastic deformation

Plastic deformation originates from the breaking of atomic bonds between neighbouring atoms and the reforming of atomic bonds between new atomic neighbours. The bonds do not return to their original configuration on removal of an applied load; the deformation is permanent. The process is greatly aided by the presence of dislocations in the metal. Higgins (1983) defines dislocations as faults or distorted regions in a crystal lattice structure, for example where a plane of atoms may not join with their nearest neighbours, forcing the surrounding planes to bend around the end of the plane, as shown in Figure 2-14.
If a shear stress is applied to the material, the dislocation moves along the direction of shear on crystallographic planes, consuming energy as the atomic bonds are broken and reformed as the dislocation moves across the lattice. As deformation progresses, the metal becomes harder and stronger as energy is consumed moving dislocations through the crystal lattice, which become entangled, resisting further deformation. This process is known as work hardening. As it becomes increasingly difficult for dislocations to move, the ductility of the metal approaches a maximum. Once maximum ductility has been reached, the material is said to be fully work-hardened (Higgins 1983).

Ramberg and Osgood (1943) developed a power-law relation to define the stress-strain locus for materials that exhibited significant levels of work-hardening. The Ramberg-Osgood relation is approximate and is shown in Equation 2-6.

$$\varepsilon_{true} = \frac{\sigma_{true}}{E} + \alpha \left(\frac{\sigma_{true}}{E}\right)^n$$

2-6

$\alpha$ is the work-hardening coefficient and $n$ is the work-hardening index. Figure 2-15 shows how different values of $n$ account for the work-hardening characteristics of a material. For example, $n = 1$ would be used for a material which behaves only in a linear elastic fashion; $n = \infty$ represents a material which is said to be perfectly plastic. The work-hardening behaviour of an austenitic stainless steel may take values between $n = 5$ to $n = 10$ (Demaid 2004).
Once the work-hardening capacity of a material has been exhausted, the tensile strength of the material, $\sigma_{TS}$, has been reached. At this point, any increase in load will not be supported by the material. Typically stainless steels have yield strengths in the range of 250 – 300 MPa and tensile strengths in the range of 570 - 650 MPa. Elongation at fracture and reduction in area can range from 40 to 60% (Higgins 1983).

### 2.4.2 The yield criterion

In a tensile test, the yield criterion is met when the stress in the specimen equals the yield strength of the material. The applied stress in a tensile test is uniaxial; it is therefore straightforward to identify the onset of plastic deformation. However, in many structures, stress acts in more than one direction. Therefore identifying the yield point, and thus the onset of plasticity, requires consideration of the effective stress in a component.

For three-dimensional bodies, the stress space has six dimensions, made up of the principal stress components in the orthogonal planes, $\sigma_{11}$, $\sigma_{22}$, and $\sigma_{33}$ and their three respective shear strains, $\sigma_{12}$, $\sigma_{13}$, and $\sigma_{23}$ (Phillips & Sierakowski 1964). If planes of principal stress are chosen so that no shear stresses act on those planes, the effective stress – the von Mises stress, $\sigma_{VM}$ – is defined as Equation 2-7 (Broek 1986).
Yield occurs when \( \sigma_{VM} = \sigma_y \). Under highly triaxial stress states, when the principal stresses are approximately equal, the von Mises stress is low, because the shear stresses are low. When the principal stresses are unequal, the von Mises stress rises, due to the presence of shear stresses, which are required to cause plastic deformation.

In three dimensions the von Mises yield criterion can be represented as the surface of a cylinder. The axis of the cylinder lies on a line where \( \sigma_{11} = \sigma_{22} = \sigma_{33} \) (Owen & Hinton 1980), illustrated in Figure 2-16.

\[
\sigma_{VM} = \frac{1}{\sqrt{2}} \left( (\sigma_{11} - \sigma_{22})^2 + (\sigma_{22} - \sigma_{33})^2 + (\sigma_{33} - \sigma_{11})^2 \right)^{1/2}
\]

Figure 2-16 - The von Mises yield surface in three dimensional stress space (Owen & Hinton 1980).

Alternatively the yield surface can be represented in two dimensions on the \( \pi \)-plane of Figure 2-16. This is shown in Figure 2-17.
To determine the proximity to yield of a given stress state, the principal stresses, and the relationship between them, must be understood. A pure hydrostatic stress, $\sigma_H$, raises or lowers each of the principal stresses by the same amount. Under such circumstances the von Mises stress is zero, and the stress state would lie on the axis of the cylinder. Yield would not occur because the yield surface has not been met.

For a given stress state in a material, the hydrostatic component is calculated by taking the mean of the applied principal stresses, Equation 2-8.

$$\sigma_H = \frac{1}{3} (\sigma_{11} + \sigma_{22} + \sigma_{33})$$  \hspace{1cm} 2-8

The proximity of a stress state to the yield surface is calculated by subtracting the hydrostatic stress component from each of the principal stress components. This defines the parameter $S$, the deviatoric stress, Equation 2-9.

$$S_{11} = \sigma_{11} - \sigma_H$$
$$S_{22} = \sigma_{22} - \sigma_H$$
$$S_{33} = \sigma_{33} - \sigma_H$$  \hspace{1cm} 2-9

The degree of triaxiality, $t$, of a stress state is defined as the ratio of the hydrostatic stress to the von Mises stress, Equation 2-10 (Demaid 2004).
In the case of real components, with finite dimensions and free surfaces, the von Mises yield surface is breached due to one or more of the principal stress components having significantly different values to the others. The onset of plasticity can be calculated through knowledge of the component geometry and the applied stress state, which is discussed in greater detail in the following section.

2.4.3 Summary

The plastic behaviour of a material can be characterised by Ramberg-Osgood parameters, which can be used to predict the response of materials to applied stresses. Plastic deformation of a material cannot always be assumed to occur on reaching the material yield stress; the applied stress state in three dimensions, which is related to the geometry of a component, has to be taken into consideration.
2.5 Fracture Mechanics

This section introduces the principles behind fracture mechanics in the elastic and elastic-plastic regimes, and how the fracture toughness of engineering materials can be characterised.

2.5.1 Linear Elastic Fracture Mechanics

Consider a metal plate under tension, causing a stress, $\sigma$, in the plate. Lines of equal stress run through the plate in the same direction as the applied load, illustrated in Figure 2-18.

Postulating an edge crack of length $a$ in the plate, as shown on the right in Figure 2-18, the lines of stress are forced to move around the crack. This has the effect of concentrating the stress close to the crack tip. The stress acts perpendicular to the plane of the crack, known as Mode I loading. The three principal loading modes to which a component may be subjected are shown in Figure 2-19. Demaid (2004) explains that in homogenous metals, cracks generally grow in accordance with the largest opening loads, and therefore Mode I is considered to be the most severe.
The stress close to the crack tip is much greater than the stress towards the back face of the plate, where the lines of stress remain unchanged. To describe the stress distribution ahead of the crack tip in an elastic material, Irwin (1948) used functions originally developed by Westergaard (1939), to define a new parameter, the Stress Intensity Factor (SIF), $K$, to relate the concentration of stress at a crack tip to an applied stress. The geometry used to define the crack tip stress fields is shown in Figure 2-20.

Figure 2-19 - The three modes of loading (Milne et al. 1988).

![Figure 2-19](image)

Mode I: Tension

Mode II: In-Plane Shear

Mode III: Out-of-Plane Shear

Figure 2-20 - Cartesian and polar co-ordinates used to define the stress state at a point $P$ ahead of a crack tip.

![Figure 2-20](image)
Equation 2-11 expresses the relationship between the SIF and the stress at the point \( P \) at a distance \( r \) ahead of a crack tip.

\[
\begin{align*}
\sigma_{xx} &= \frac{K_I}{\sqrt{2\pi r}} \cos\left(\frac{\theta}{2}\right) \left[1 - \sin\left(\frac{\theta}{2}\right) \sin\left(\frac{3\theta}{2}\right)\right] \\
\sigma_{yy} &= \frac{K_I}{\sqrt{2\pi r}} \cos\left(\frac{\theta}{2}\right) \left[1 + \sin\left(\frac{\theta}{2}\right) \sin\left(\frac{3\theta}{2}\right)\right] \\
\sigma_{zz} &= \nu(\sigma_{xx} - \sigma_{yy}) \\
\tau_{xy} &= \frac{K_I}{\sqrt{2\pi r}} \sin\left(\frac{\theta}{2}\right) \cos\left(\frac{\theta}{2}\right) \cos\left(\frac{3\theta}{2}\right)
\end{align*}
\]

Considering the equations defining \( \sigma_{xx} \) and \( \sigma_{yy} \) in Equation 2-11, the stresses change in proportion to the inverse of the square root of the distance from the crack tip. Considering the stress directly ahead of the crack tip – and thus removing the angular variation in Equation 2-11 – the opening stress can be expressed by Equation 2-12 (Demaid 2004).

\[
\sigma_{yy} = \frac{K_I}{\sqrt{2\pi r}}
\]

Equation 2-12 is plotted in Figure 2-21, which shows that the stress ahead of a crack increases rapidly as the crack tip is approached, reaching a singularity at the crack tip.

![Figure 2-21](image)

**Figure 2-21** – Variation of crack opening stress with increasing distance from the crack tip in an edge-cracked plate with an applied \( K = 75 \text{ MPa}\sqrt{\text{m}} \).
To evaluate $K_f$ fully a geometry parameter, $Y$, is also required to describe the concentration of stress at the tip of a crack. Solutions for $Y$ are provided in handbooks and software packages for common crack and component geometries (Demaid 2004).

Equation 2-13 relates $K_f$ in a cracked component under an applied stress of $\sigma$ and crack length $a$. The units of $K$ are expressed in MN m$^{3/2}$ or MPa√m.

$$K_f = Y \sigma \sqrt{\pi a}$$ \hspace{1cm} 2-13

$K_f$ is suitable for characterising fracture under linear elastic conditions. The singularity in Figure 2-21 shows that at the crack tip, the stresses are assumed to approach infinity, well in excess of the yield stress. In reality, a plastic zone is formed ahead of the crack. If the plastic zone is small compared to the overall size of the component, the elastic response of the material to the applied stress dominates; $K$ is a suitable crack driving force parameter. However, if the plastic zone is large, causing an elastic-plastic material response, $K$ is no longer a suitable crack driving force parameter.

The plastic zone size, $r_p$, can be approximated using the Westergaard functions (Broek 1986). Equation 2-14 is the plastic zone size for plane strain conditions and Equation 2-15 for plane stress conditions. The concepts of plane strain and plane stress are detailed in Section 2.5.3.

$$r_p = \frac{1}{6\pi} \left( \frac{K_f}{\sigma_y} \right)^2$$ \hspace{1cm} 2-14

$$r_p = \frac{1}{2\pi} \left( \frac{K_f}{\sigma_y} \right)^2$$ \hspace{1cm} 2-15
2.5.2 Elastic Plastic Fracture Mechanics

In work-hardening materials, where extensive plasticity can develop ahead of a crack tip, the stress fields described by the Westergaard functions no longer apply. In the elastic-plastic regime, the stress fields are characterised by Hutchinson, Rice and Rosengren (HRR) fields (Hutchinson 1968; Rice & Rosengren 1968). The component of stress, in polar co-ordinates, is expressed in Equation 2-16.

\[
\sigma_{ij} = \sigma_y \left( \frac{EJ}{a\sigma_y^2 l_n r} \right)^{\frac{1}{n+1}} f_{ij}(\theta, n)
\]

Where, \(l_n\), an integration parameter, is dependent on the Ramberg-Osgood work-hardening index, \(n\). \(r\) is the distance from the crack tip, \(\alpha\) is the work-hardening coefficient, \(E\) is Young’s Modulus and \(f_{ij}(\theta, n)\) is a function of \(n\) and the polar coordinate \(\theta\).

The parameter \(J\) was developed by Rice (1968), to describe the strain energy distribution around the crack tip. \(J\) is defined in Equation 2-17, where \(U\) is mechanical potential energy and \(a\) is the crack length.

\[
J = -\left( \frac{\partial U}{\partial a} \right)
\]

\(J\) can be evaluated as a path-independent line integral, Equation 2-18.

\[
J = \int_{\Gamma} \left( Wdy - T \frac{\partial u}{\partial x} dS \right)
\]

The parameters in Equation 2-18 are illustrated in Figure 2-22. The path along which \(J\) is evaluated is denoted by \(\Gamma\), \(W\) is the strain energy density, \(T\) is a traction vector and \(u\) is the corresponding displacement vector.
Demaid (2004) describes $J$ as the rate of change of work that is put into a material as a function of crack growth. Its units are kJ/m$^2$. In the elastic-plastic fracture mechanics (EPFM) regime, $J$ characterises the strength of the stress-strain field at a crack tip.

Under Small-Scale Yielding (SSY) conditions, where the size of the plastic zone is relatively small compared to the surrounding body of elastic material, $J$ can be related to $K$, through the relation in Equation 2-19, where $K_j$ is the linear-elastic SIF calculated from the elastic-plastic $J$. $K$ can be considered a special case of $J$ operating in the elastic regime (Demaid 2004).

$$K_j = \frac{EJ}{\sqrt{(1 - \nu^2)}}$$  \hspace{1cm} 2-19

Under conditions when plasticity is extensive, neither LEFM nor EPFM accurately describes the crack tip stress fields. Cases where large strain effects take place at the crack tip, such as extensive crack tip blunting, remove the HRR singularity. Under cases of large scale yielding, for example, where plasticity reaches from the crack tip to the back face of a component, plastic collapse solutions are used. Demaid et al. (2009) explain that plastic collapse precludes the failure of a material by fracture, occurring as net section yielding or buckling. Plastic collapse is, essentially, the opposite of brittle fracture.
Smith (1992) explains that the analytic expressions which describe the stress-strain fields at the crack tip require the use of numerical methods to calculate $J$ in ductile components containing defects. FEA is often used under such circumstances (Scheider & Brocks 2008). This can be a time-consuming approach, which has only become a practical option in recent years with the advent of powerful desktop computers. Handbook $J$-solutions have been created for common components and defect geometries, such as the Electric Power Research Institute (EPRI) $J$-estimation handbook (Kumar et al. 1981).

### 2.5.3 Material Fracture Toughness

When performing structural integrity assessments, it is necessary to know the fracture toughness of the material from which a structure is made. Fracture toughness is a measure of the ability of a material to resist crack extension (Sherry & Marrow 2010). To make conservative structural integrity assessments, a lower-bound measure of fracture toughness is required.

The plastic zone ahead of a crack absorbs the energy of fracture; plasticity causes the applied stresses to redistribute. Therefore more work has to be done for the crack to grow through the plastic zone. Demaid (2004) explains that the toughness of a metal arises from the need of a growing crack to overcome plasticity. Therefore large plastic zones consume more energy than small plastic zones: higher loads are required to cause fracture. For a lower-bound fracture toughness measurement, test specimens have been designed to ensure that the plastic zone is small compared to the surrounding body of elastic material.

Consider the plate with the edge crack of length $a$ illustrated in Figure 2-18. The cross-section of the plate, shown in plan view, is shown in Figure 2-23.
The plastically deformed material directly ahead of the crack tip, in the plane of the crack, is shown as the shaded area in Figure 2-23, the shape of which is influenced by the geometry of the plate. This plastic zone is described in more detail in Figure 2-24.

Deep into the body of the plate, the zone ahead of the crack tip experiences very high stresses, in excess of yield. However, a lot of material surrounds this area in all three dimensions; the stress state is highly triaxial. The surrounding material constrains the material ahead of the crack tip from yielding; only the highest stresses, closest to the crack tip, can cause yielding. These are plane strain conditions; the strain component is constant or zero perpendicular to the plane of
the plate (Callister 2007). Thus, the plastic zone is small ahead of the crack in the centre of the plate.

At the edges of the plate, conditions of plane stress exist; the stresses out of the plane of the plate are zero, because the surfaces of the plate are free to move. The level of triaxiality is low; therefore less stress is required to breach the yield surface in stress space. As a result, more material ahead of the crack tip can yield, increasing the size of the plastic zone towards the edges. Demaid (2004) illustrates the plastic zone ahead of a crack tip in three dimensions in Figure 2-25.

![Figure 2-25 - Plastic zone shape across a crack tip in a plate specimen (Demaid 2004).](image)

For a lower-bound fracture toughness measurement, the majority of the crack front must experience plane strain conditions, to keep the plastic zone small. In the LEFM regime, the fracture toughness is expressed as the value of $K_I$ at which fracture occurs as a single event; this critical condition is denoted $K_{ic}$, the plane strain fracture toughness.

Fracture toughness specimens are designed with sufficient thickness to enforce plane strain conditions across most of the crack front. For example, consider the plate in Figure 2-26a. The specimen is sufficiently thick to permit plane strain conditions along most of the crack front, accompanied by plane stress conditions at
the edges. Figure 2-26b shows a thinner plate, of the same material, containing a crack of the same length. Most of the crack front lies close to a free surface, causing plane stress conditions to dominate, permitting a large plastic zone. Therefore the thin plate will not yield a conservative, lower bound measure of fracture toughness.

![Figure 2-26 - Plastic zone sizes in two cracked plates. The material and crack lengths are the same for both specimens.](image)

Callister (2007) explains that once the thickness of a specimen becomes sufficiently large, the measured fracture toughness becomes independent of thickness, because plane strain conditions dominate the stress state along the crack tip at fracture.

The type of material under test must also be taken into account. For example, ferritic steel undergoes a transition from brittle to ductile fracture behaviour with increasing temperature, illustrated in Figure 2-27.

![Figure 2-27 – Brittle-ductile transition behaviour of ferritic steels (based on Demaid, 2004).](image)
The brittle fracture regime lies at temperatures below the transition temperature of the material, known as the “lower-shelf”. In the transition region, a mixture of ductile and brittle fracture takes place. As temperature increases, ductile failure begins to dominate, until the ductile fracture regime is reached, known as the “upper-shelf” (Demaid 2004). Austenitic steels, however, do not exhibit transition behaviour; ductile fracture behaviour is maintained at all temperatures (Higgins, 1983).

The applicability of LEFM and EPFM therefore depends on the material of interest. In ferritic steels on the lower-shelf, LEFM describes failure most accurately, because brittle fracture is accompanied by little plasticity. A small plastic zone consumes less energy, allowing fracture to take place under low loads. On reaching a critical crack driving force (CDF), fracture is sudden in nature; the fracture toughness is characterised by a single failure event through a parameter such as $K_{IC}$.

In austenitic steels and ferritic steels on the upper-shelf, ductile failure conditions exist; plasticity ahead of the crack tip is more widespread. EPFM describes the stress state ahead of the crack tip; the $J$-integral is used to characterise the CDF. On reaching a critical CDF, stable ductile tearing takes place with increasing load (Demaid et al. 2009). As a result, the critical CDF is defined through a measure of ductile tearing, using $J$-resistance ($J$-$R$) curves, produced by loading fracture toughness specimens and recording the amount of stable crack growth, $\Delta a$. Under EPFM conditions, fracture toughness specimens tend to be thicker, to constrain the plastic zone, encouraging small-scale yielding conditions, so that a conservative measure of fracture toughness is obtained. Test standards and procedures have been developed so that an appropriate measure of the fracture toughness of brittle and ductile materials can be made.

### 2.5.4 Fracture toughness test standards

Fracture toughness test standards, such as BS7448 (British Standards Institution 2005a), ASTM E1820 (American Society for Testing and Materials 2011b) and procedures such as ESIS P2-92 (European Structural Integrity Society 1992) have been developed to ensure that appropriate measurements of material fracture
toughness are made. Test specimens most commonly take the form of Single-Edge Bend, SE(B), specimens or Compact Tension, C(T), specimens. Test standards specify minimum dimensions of fracture toughness specimens, based on the flow strength of the material under test. For brittle materials or those that exhibit limited ductility, specimen dimensions are designed to enforce plane strain conditions so that a lower bound $K_{ic}$ value can be obtained. The plastic zone at the crack tip, $r_p$, is chosen to be less than approximately $1/50^{th}$ of the specimen dimensions (Sherry & Marrow 2010). For work-hardening materials, specimen dimensions are designed to limit plasticity, encouraging small-scale yielding, so that $J$-control conditions apply.

For example, BS7448 and ESIS P2-92 require the crack length, $a$, specimen thickness, $B$, width, $W$ and uncracked ligament length, $(W - a)$, to meet the condition in Equation 2-20 to obtain a valid measurement of $K_{ic}$ (British Standards Institution 2005a; European Structural Integrity Society 1992).

$$a, B, (W - a) \geq 2.5 \left( \frac{K_i}{\sigma_y} \right)^2$$  \hspace{1cm} 2-20

For more ductile materials, a validity limit, $J_{max}$, based on the dimensions of the test specimen, is imposed, to ensure that sufficient constraint is maintained in a test specimen to obtain a conservative measure of fracture toughness. $J_{max}$ is calculated from the smaller of Equation 2-21 and Equation 2-22 (European Structural Integrity Society 1992).

$$J_{max} = (W - a) \frac{\sigma_{flow}}{20}$$  \hspace{1cm} 2-21

$$J_{max} = B \frac{\sigma_{flow}}{20}$$  \hspace{1cm} 2-22

Limits are set on the amount of crack growth permitted in a fracture toughness test, $\Delta a_{max}$, defined by Equation 2-23, to ensure the plastic zone does not reach a free surface, influencing the fracture toughness measurement.
During a fracture toughness test, the load versus displacement behaviour of the specimen is recorded. Crack Mouth Opening Displacement (CMOD), or Load Line Displacement (LLD), is used to measure displacement (European Structural Integrity Society 1992). Loading rates are specified to avoid the effects of dynamic loading influencing test results, particularly in strain-rate sensitive materials (Sherry & Marrow 2010).

For $K$-control fracture, $K_{IC}$ is calculated using Equation 2-24 (European Structural Integrity Society 1992).

$$K_{IC} = \frac{F_c}{B\sqrt{W}} f \left( \frac{a}{W} \right)$$  \hspace{1cm} 2-24

$F_c$ is the load at fracture and $f \left( \frac{a}{W} \right)$ is a geometry function, dependent on the use of an SE(B) specimen or a C(T) specimen.

Under $J$-control, $J$ is calculated according to Equation 2-25. $U$ is the area under the force-displacement curve and $\eta$ is a geometry function, equal to 2 for SE(B) specimens (European Structural Integrity Society 1992).

$$J = \frac{\eta U}{B(W - a)}$$  \hspace{1cm} 2-25

$J$ versus crack extension, $\Delta a$, is plotted to produce a $J$-resistance curve. An example is shown in Figure 2-28. The choice of a critical value of $J$ lies with the tester, often to conform to national, international or industry-wide conventions.
One critical measure is $J_{\text{initiation}}$, the CDF at the point of tearing just after crack blunting has taken place. From a practical standpoint, this parameter is hard to judge in the engineering laboratory without the use of a Scanning Electron Microscope (Østby et al. 2007). More conventionally, the CDF which corresponds to 0.2 mm of ductile tearing is used, $J_{0.2}$. Alternatively, a 0.2 mm offset from the blunting line can be made. The intercept of this line with the $J$-$R$ curve yields an alternative measure of toughness, $J_{0.2BL}$ (Demaid et al. 2009).

### 2.5.5 Summary

Fracture mechanics brings together the concepts of elasticity and plasticity so that the response to loading of materials containing defects can be predicted and understood. This has resulted in the development of test methodologies to ensure that the fracture behaviour of materials can be quantified accurately. A review of $J$-resistance curve behaviour in stainless steel pipe welds is given in Section 2.8.3.
2.6 Ductile fracture

2.6.1 Ductile fracture mechanism

Austenitic stainless steels are ductile in nature. Fracture in ductile materials can be viewed as a three stage process, illustrated in Figure 2-29: void initiation, growth and coalescence. The first stage centres on hard second phase particles, such as MnS inclusions. When the local plastic strain reaches sufficiently high levels, the second phase particles can either crack or debond from the matrix, initiating microvoids at these sites (Taktak et al. 2009). If the second phase particles are part of a tightly bound matrix, this first stage may lead straight to fracture (Rakin et al. 2000). In materials which exhibit distinct plastic behaviour, such as austenitic stainless steels and aluminium alloys, a second stage follows: the high stress triaxiality and high levels of plastic strain act together at the sites of the second phase particles to cause microvoids to grow. The third stage of the fracture process can take place in two ways: the material between voids may fail along microscopic shear bands. Alternatively, a second distribution of finer voids at nano-precipitates, such as carbides, initiate, causing the voids to coalesce. Ductile damage commences at a slow rate, accelerating to a much higher rate as fracture initiation is approached (L. Xue & Wierzbicki 2008).

![Figure 2-29 - The three stages of ductile fracture, a) initiation of a void at a second phase particle, b) growth of void and initiation of new voids at other second phase particles and c) coalescence of voids and thus crack growth (Sherry et al. 2008).](image-url)
Tvergaard and Niordson (2008) explain that crack growth can take place in two different ways. For materials with a low void volume fraction, a crack tip will blunt and interact with the first void ahead of the crack tip. Once the void and crack tip have coalesced, the new crack tip will begin to interact with the next void and so forth. This is known as the void-by-void mechanism. However, if the initial void volume fraction is sufficiently high, several voids on the same plane ahead of the crack tip can interact with each other at the same time, and can initiate and grow together. This is known as the multiple void mechanism.

Energy is consumed throughout the fracture process. Rakin et al. (2000) describe how materials which exhibit plastic behaviour will generally undergo all three stages of ductile fracture, but warn that it is not straightforward to determine whether any given material will behave in a ductile or brittle fashion. This is because material behaviour will depend on factors such as its fabrication history, condition, loading geometry and temperature.

### 2.6.2 Ductile fracture in stainless steel

Stainless steel generally exhibits all three stages of ductile fracture. Voids initiate at sites of inclusions, such as MnS, which debond from the material matrix at the early stages of fracture. With increasing stress, the voids grow and begin to coalesce. New voids begin to initiate on second phase particles, such as carbides. The smaller voids also begin to grow and coalesce, until the ligaments between the network of voids become plastically unstable and fracture. This is shown in an SEM image of the fracture surface of a sample of 304LN stainless steel, by Das et al. (2008), Figure 2-30. The large voids initiated first around inclusions; the material between these large voids contains several smaller voids, which initiated around smaller carbides at the later stages of fracture.
A more detailed image of a stainless steel fracture surface is shown in Figure 2-31, by Herms et al. (1999), of a 316L stainless steel. A large void, initiated at the early stage of fracture, has coalesced with smaller neighbouring voids, which initiated later in the fracture process.
2.7 Structural integrity assessment

Structural integrity assessment procedures have been developed to aid engineers in measuring and predicting the behaviour of structures under design conditions, which includes assessing the integrity of structures containing defects. Extensive research and development has led to the development of assessment procedures such as BS7910 (British Standards Institution 2005b), API RP-579 (American Petroleum Institute 2000) and RSE-M (Faidy 2000), based on the requirements and practices from the respective industries in which the procedures originated. This section focuses on the R6 procedure, originally developed by the UK Central Electricity Generating Board (CEGB) to assess structures in the UK nuclear power generation industry (British Energy Generation Limited 2000).

2.7.1 The R6 defect assessment procedure

The R6 procedure provides advice on assessing the integrity of structures containing defects. It was originally developed for use in the UK nuclear power industry, but since then its principles have been used in other industries and standards bodies, such as the British Standards Institution’s BS7910 procedure (British Standards Institution 2005b).

R6 was introduced in 1976 and has been continuously developed ever since; it is now in its fourth revision, having been applied to both nuclear and conventional electricity generating plant (Budden et al. 2000). The R6 development programme keeps the procedure up to date with advances in the field of fracture mechanics so that assessments may be improved over time, reducing pessimisms when analyses are performed. For example, a review by the UK Technical Advisory Group on the Structural Integrity of Nuclear Plant (TAGSI) acknowledged that of all the assessment procedures in existence, only R6 has improved the advice given on treating secondary stresses (Dolby et al. 2008).

The R6 procedure shows the proximity to failure of a component through a straightforward graphical representation of material behaviour. It accounts for material properties in the LEFM regime, the EPFM regime and the effects of work hardening on fracture toughness.
Performing a full elastic-plastic evaluation of a cracked component can be a time-consuming and complex process. Obtaining the elastic-plastic crack driving force, $J$, through numerical methods or FEA is not always a practical means of undertaking routine assessments. Demaid et al. (2009) describe how Ainsworth developed an approach that uses ordinary stress-strain data to calculate the elastic-plastic CDF in a cracked body.

Ainsworth (1984) explains that the stress-strain curve for strain-hardening materials fits poorly to the Ramberg-Osgood power-law curve, particularly at high strains. To address this problem, Ainsworth introduced the reference stress, $\sigma_{ref}$, defined in Equation 2-26. $P$ is the load applied to the structure and $P_y$ is the collapse load for an elastic-perfectly plastic material, having a yield strength of $\sigma_y$.

$$\sigma_{ref} = \frac{P}{P_y}\sigma_y$$  \hspace{1cm} 2-26

The reference strain, $\varepsilon_{ref}$, is defined as the true strain corresponding to the reference stress on the true stress-true strain curve. The reference stress approach removes the power-law dependence of $J$ on the material hardening index, $n$. Demaid et al. (2009) show that the plastic CDF, $J_{pl}$, can be obtained using simple parameters, shown in Equation 2-27.

$$J_{pl} = \frac{K^2}{E'} \left( \frac{E \varepsilon_{ref}}{\sigma_{ref}} - 1 \right)$$  \hspace{1cm} 2-27

The elastic component of CDF, $J_{el}$ or $K$, is obtained using the standard LEFM approach described earlier. Thus the elastic-plastic CDF can be obtained to a good approximation. Ainsworth (1986) explains that $\sigma_{ref}$ corresponds to an effective primary stress giving the same $J$ as combined primary and secondary stresses.

The reference stress approach enables the elastic-plastic CDF to be calculated using ordinary stress-strain data for a particular material. The CDF as a function of
applied load can be plotted to provide a visual representation of the proximity of a cracked component to failure.

The R6 procedure takes this approach one step further, through the Failure Assessment Diagram, or FAD. Normalising the applied load, \( P \), by the collapse load, \( P_y \), creates the normalised load parameter \( L_r \), Equation 2-28.

\[
L_r = \frac{P}{P_y}
\]

Normalising the CDF produces the \( K_r \) term, Equation 2-29.

\[
K_r = \frac{K_{\text{elastic}}}{K_{\text{elastic-plastic}}}
\]

Normalising the applied load and CDF parameters allows for the CDF to be plotted as a function of load for a range of materials on the same graph. Demaid et al. (2009) show how Ainsworth’s reference stress approach is used to produce a Failure Assessment Curve (FAC). The curve is defined by the function \( f(L_r) \), Equation 2-30.

\[
f(L_r) = \left[ \frac{E \varepsilon_{\text{ref}}}{L_r \sigma_y} + \frac{L_r^3 \sigma_y}{2E \varepsilon_{\text{ref}}} \right]^{-1/2}
\]

The FAC is material-specific; a range of FACs can be displayed on a single FAD. After performing analyses on a range of components made of different materials and containing a range of crack geometries, a lower-bound FAC was created, \( f_1(L_r) \), whose formula is shown in Equation 2-31 (British Energy Generation Limited 2000).

\[
f_1(L_r) = \frac{0.3 + 0.7 \exp(-0.6L_r^6)}{(1 + 0.5L_r^2)^{1/2}}
\]

This is known as the Option 1 FAC. The material-specific FAC described by Equation 2-30 is the Option 2 FAC, which, for clarity, is henceforth defined as \( f_2(L_r) \). An example is shown in Figure 2-32. Option 1 is so-called as it is the first approach in
assessing a component containing a defect. Option 1 analyses require knowledge of the yield strength, tensile strength, fracture toughness and basic geometry of a component to make an assessment. Option 2 is a more sophisticated approach, which requires further information in the form of stress-strain data for the material under assessment.

The FAD is essentially a $J$-estimation scheme, able to show the proximity of a component to failure without an assessor having to resort to complex elastic-plastic analyses (Demaid 2004). To complete the FAD, a cut-off, $L_r^{max}$, is applied to the FAC to prevent an assessor from making use of work-hardening capacity that may not be available in a material. Strongly work-hardening materials can exhibit values as high as $L_r^{max} = 1.8$. $L_r^{max}$ is calculated by Equation 2-32 (British Energy Generation Limited 2000).

$$L_r^{max} = \frac{\sigma_{flow}}{\sigma_{yield}} \quad 2-32$$

Where $\sigma_{flow}$ is the flow stress, defined in Equation 2-33.

$$\sigma_{flow} = \frac{\sigma_y + \sigma_{TS}}{2} \quad 2-33$$

![Figure 2-32 - Option 1 and Option 2 FAC for 304-stainless steel at room temperature - material data from Goldthorpe (2011).](image)
### 2.7.3 Failure Assessment Point

The Failure Assessment Point (FAP) is used to locate a component under assessment on the FAD, from which its proximity to failure can be evaluated. The FAP, with co-ordinates \((L_r, K_r)\), requires the fracture toughness, \(K_{mat}\), of the material to be known. For a FAP, \(K_r\) is obtained using the relation in Equation 2-34; Equation 2-28 is used to obtain \(L_r\).

\[
K_r = \frac{K_{\text{elastic}}}{K_{\text{mat}}}
\]  

2-34

If the FAP lies in the region bounded by the curve, the component has not reached a critical condition. If the FAP lies outside the FAC, then the component is deemed unsafe (Demaid et al. 2009).

The FAD covers the whole spectrum of failure, as illustrated in Figure 2-33. At low \(L_r\), failure is accompanied by limited plasticity, operating in the LEFM domain, region A. As \(L_r\) increases, plasticity becomes more widespread, region B. Beyond approximately \(L_r = 1.05\) plasticity has become so extensive that plastic collapse becomes the failure mode, region C (Ainsworth 1986).

![Figure 2-33 - Option 1 Failure Assessment Curve, covering the three primary failure modes.](image-url)
An advantage of using the FAD approach is that it enables an assessor to see the effect of a change of input parameter on the resultant assessment, in a clear and straightforward fashion. Figure 2-34 shows the effect that a change in load, crack length, material yield strength, fracture toughness or flow strength may have on an assessment point. For example, if the load is increased, \( L_r \) and \( K_r \) will increase in proportion, moving the FAP towards the FAC. If a material exhibits a higher toughness than previously measured, the FAP correspondingly moves vertically downwards, away from the FAC.

![Graph showing the effect of changing parameters on a FAP](image)

Figure 2-34 - The effect of changing parameters, such as crack length, yield strength and fracture toughness of a material on a FAP, based on Demaid (2004).
The FAD also allows for a reserve factor, $F$, to be easily calculated, as shown in Figure 2-35 (Milne et al. 1988).

The reserve factor is defined using Figure 2-35 and Equation 2-35. Point $O$ is the origin, Point $A$ is the assessment point and Point $B$ is where the FAP will meet the FAC under increased applied load.

$$ F = \frac{OB}{OA} \quad \text{(2-35)} $$

When assessing a component containing a defect, the Option 1 analysis is usually applied first. Should the FAP lie close to the FAC, an Option 2 analysis may be used. For example, in Figure 2-32, a FAP lying close to the Option 1 FAC at $L_r = 0.95$ lies well within the region bounded by the Option 2 FAC. In this case, the ductility of 304 stainless steel permits greater margins of safety under such loads. The Option 2 FAC is less conservative than the Option 1 FAC for this ductile material, but requires more work on the part of the assessor to allow such an assessment to be made, requiring analysis of the tensile data for the material under scrutiny.
R6 also offers a third approach to producing a FAC, the Option 3 curve. Budden et al. (2000) explain that an Option 3 FAC requires a full elastic-plastic analysis, which is usually undertaken through numerical modelling approaches. As a result, the material properties, crack size and geometry of the component must be known before an Option 3 analysis can be performed. The elastic-plastic CDF, $J_{el-pl}$, is often obtained using FEA, and converted to an equivalent LEFM parameter, $K_f$, using Equation 2-36.

$$K_f = \sqrt{\frac{E J_{el-pl}}{(1 - v^2)}} \quad 2-36$$

The Option 3 FAC, $f_3(L_r)$, is defined by Equation 2-37, the elastic CDF divided by the elastic-plastic CDF.

$$f_3(L_r) = \frac{K_i}{K_f} \quad 2-37$$

Mattar-Neto et al. (2008) describe how Option 1 is best applied to situations where brittle fracture is the failure mode, with little or no ductility present. Option 2 can deal with conditions where greater ductility may be encountered. For high toughness materials, where assessing the tearing toughness is of interest, Options 2 and 3 will provide a more accurate analysis than Option 1, as the effects due to extensive plasticity and work hardening can be accounted for better than would be possible through an Option 1 analysis.

### 2.7.4 Residual stress in R6

#### 2.7.4.1 Classification of primary and secondary loading

Blackburn (1986) highlighted the need to account for secondary loading in assessments carefully, as, at the time, treating all loads on a component as primary was leading to pessimistic predictions of failure. The simplest approach involved treating all loads as primary and used a FAD to determine proximity to failure. The most accurate approach required full non-linear FEA.
Banahan (2008) and Hallbäck and Nilsson (1992) explain that classifying stresses as primary or secondary is not always a straightforward process. Generally, stresses arising from mismatch – such as thermal gradients and weld residual stresses – are classified as secondary, and are often self-equilibrating. However, some thermal and residual stresses which self-balance across an entire structure will not necessarily balance locally across a flaw, and therefore, may not be treated as secondary stresses (Banahan 2008).

R6 advises that if in doubt over the classification of a stress, it should be treated as primary. However, this may lead to an overly conservative result, predicting a low safety margin. As Dolby et al. (2008) point out, experience from performing real assessments showed that these predicted low-margins of safety were not consistent with practical experience. Banahan (2008) explains that as the level of analysis is raised (e.g., from Option 1 to Option 2), the assessment of residual stress becomes less conservative.

2.7.4.2 Combined loading

A secondary stress, $\sigma^S$, such as a weld residual stress field, will contribute to the crack tip stress fields in a component containing a defect. The elastic SIF, $K_f^S$, can be calculated using Equation 2-38, analogous to Equation 2-13.

$$K_f^S = Y_\sigma^{S/\pi a}$$  \hspace{1cm} 2-38

Under combined loading – where both primary and secondary stresses act on a component - $K_f^S$ can be added to the primary SIF, $K_f^P$, to obtain a value for $K_f$. With no primary load applied, $K_f = K_f^S$. With increasing applied primary load, $K_f$ increases proportionally, as shown in Figure 2-36.
Where secondary stresses exist, contributing to the total CDF, the FAD in Figure 2-36 suggests that a critical condition is reached at a lower value of $L_r$ than would be the case for primary loading only. It also suggests that the failure may be more brittle in nature under combined loading. However, experience of operating real components, especially those made from strain-hardening materials, has shown that such predictions are often over-conservative. This is because the effect of plasticity on secondary stresses has not been fully taken into consideration.

Ainsworth (1986) introduced an interaction parameter, $\rho$, to account for the way in which primary and secondary stresses interact, using the reference stress approach. The reference stress can be related to the $L_r$ parameter, by rearranging Equation 2-26 to produce Equation 2-39. $\sigma_{\text{ref}}$ can be considered to be the effective stress in a cracked component, causing the same CDF as combined primary and secondary stresses.

\[
L_r = \frac{\sigma_{\text{ref}}}{\sigma_y}
\]

2-39

Therefore $\sigma_{\text{ref}}$ is dependent on the reference stress due to primary loading, $\sigma_{\text{ref}}^p$, and the reference stress due to secondary loading, $\sigma_{\text{ref}}^s$. Ainsworth (1986) derived a relationship between $\sigma_{\text{ref}}^p$ and $\sigma_{\text{ref}}^s$ to show how the overall reference stress was...
influenced by a combination of primary and secondary stresses. Figure 2-37 shows how increasing $\sigma_{ref}^p$ changes $\sigma_{ref}$ when different levels of $\sigma_{ref}^s$ are present. All of the terms are normalised by the yield stress, $\sigma_y$.

![Figure 2-37 - Reference stress under combined loading (Hooton et al. 2006).](image)

Figure 2-37 shows that when no secondary stress is present, $\sigma_{ref}^p = \sigma_{ref}$. For increasing $\sigma_{ref}^s$, $\sigma_{ref}$ becomes enhanced by the combined primary and secondary loads at low primary loads. The reference stress then begins to reduce with increasing primary load, as the contribution from the secondary stresses diminishes due to residual strains being overwhelmed by plasticity (Ainsworth 1986).

Applying this approach to the R6 procedure, Ainsworth (1986) described the effects of combined primary and secondary stresses through a shift in the $K_r$ parameter, denoted $\psi$. Figure 2-38 shows $\psi$ plotted as a function of $L_r$ for different levels of secondary stress.
At low (primary) mechanical loads, $\psi$ increases strongly up to a maximum, reflecting the enhancement of $K_r$ caused by the interaction between the primary and secondary stresses. As the primary load increases, plasticity develops, redistributing the secondary stresses. This is shown by $\psi$ falling to zero at net section yield, where $\sigma_{ref} = \sigma_y$. Beyond this point, $\psi$ becomes negative, where plasticity corrections to the total SIF are not necessary (Ainsworth 1986). This forms the basis for defining the $\rho$-parameter in the R6 procedure; the simplified procedure takes an idealised form of Figure 2-38, shown in Figure 2-39.

Figure 2-38 - Variation of the shift of $\psi$ with primary and secondary loads (Hooton et al. 2006).

Figure 2-39 – The $\rho_1$ parameter as defined in R6 Revision 4 (British Energy Generation Limited 2000). This is an idealised curve encompassing the curves in Figure 2-38 to remain conservative.
An alternative plasticity correction factor, $V$, was incorporated into Revision 4 of R6. Both $\rho$ and $V$ approaches are equivalent (British Energy Generation Limited 2000), although Goldthorpe (2000) acknowledges the $V$ factor as a simpler way of accounting for the interaction between stresses due to primary and secondary loading.

2.7.4.3 The simplified procedure

The linear elastic SIF, $K_i^S$, is calculated based on the flaw size and the secondary stress, $\sigma^S$, acting on the flaw, using Equation 2-13. $K_i^S$ is positive for crack-opening stresses. If $K_i^S$ is negative or zero, $\rho$ and $V$ are set to zero and 1 respectively, and no further steps need to be taken (British Energy Generation Limited 2000).

The ratio of the primary SIF to the normalised load parameter, $K_i^p / L_r$, is evaluated. This is followed by evaluating the ratio of $K_i^S / (K_i^p / L_r)$, the ratio of secondary crack driving force to the primary crack driving force as a proportion of the applied primary load.

To evaluate $\rho$, the following steps are taken:

- The parameter $\rho_1$ is evaluated from a plot of $\rho_1$ versus $K_i^S / (K_i^p / L_r)$, shown in Figure 2-39. The value calculated for $K_i^S / (K_i^p / L_r)$ in the previous step is used to locate a point on the curve; $\rho_1$ is the corresponding value on the vertical axis.
- The $\rho$-parameter is defined as follows.

\[
\begin{align*}
\rho &= \rho_1, & L_r \leq 0.8 \\
\rho &= 4\rho_1 (1.05 - L_r), & 0.8 < L_r \leq 1.05 \\
\rho &= 0, & 1.05 < L_r.
\end{align*}
\]
Alternatively, to evaluate $V$, the following applies:

\[
V = 1 + 0.2L_r + 0.02 \frac{K_f^S}{K_f^P/L_r} (1 + 2L_r), \quad L_r < L_r^c
\]

\[
V = 3.1 - 2L_r, \quad L_r^c < L_r < 1.05
\]

\[
V = 1, \quad L_r > 1.05.
\]

Where $L_r^c < 1$ is a function of $K_f^S/(K_f^P/L_r)$, evaluated by the intersection of the formulae on the first two lines of Equation 2.41 (British Energy Generation Limited 2000).

2.7.4.4 Limitations of the interaction parameter

Budden et al. (2000) explain that the simplified method causes the $\rho$-parameter to be over-conservative where extensive plasticity causes redistribution of secondary stresses. The detailed approach must be used instead; this permits the $\rho$-parameter to be negative, so that the redistribution of secondary stresses can be accounted for.

Prior to Revision 4 of the R6 procedure, containing the detailed method, alternative $\rho$-parameters were proposed by several authors. For example, Qi (1992) introduced the parameters $\rho_2$ and $\rho_3$ for the R6 Option 2 and Option 3 curves respectively, the aim of which was to better reflect the contribution of residual stress in the deformation of structures containing weld defects. Pan and Li (1997) highlighted a situation where the $\rho$-parameter led to a non-conservative evaluation of $K_r$ in a component containing a small crack in a high tensile uniform secondary stress field. Pan and Li (1997) presented a new $\rho$-parameter for pure secondary loads on the R6 Option 1 FAD to address the situation of a small crack in a high tensile stress field. This involved evaluating the work of Qi (1992), whose $\rho$ factor was deemed over conservative by Pan and Li, setting their own values of $\rho$ between Qi’s results and those of R6. This led to a failure assessment point for a small crack in a high uniform tensile stress field to lie just above the Failure Assessment Curve, reflecting the susceptibility to failure of a structure better than either Qi’s method or that of R6.
The limitations of the $\rho$-parameter have been better accounted for since Revision 4 of R6 was introduced (British Energy Generation Limited 2000), which states that the simplified procedure should not be used under the following circumstances:

- When only secondary stresses are present, as $K_f^p$ is undefined,
- When secondary stresses are large, as $\rho$ will be over-conservative,
- When secondary loads are combined with small primary loads, because $\rho$ does not tend to zero in the limit of negligible primary loads,
- If elastic follow up of the material is judged to be significant, for example, if a small defect is present in a structure many times larger than the defect itself.

Under such circumstances, the detailed procedure may be used to evaluate a suitable plasticity correction term.

### 2.7.4.5 The detailed procedure

R6 Revision 4 (British Energy Generation Limited 2000) states that the detailed procedure should be applied when $K_f^S/(K_f^p/L_r) > 4$, when secondary stresses act alone, or when applied primary stresses are low.

First, the linear elastic SIF, $K_f^S$, is calculated. The effective SIF, $K_f^e$, is also calculated. As with the simplified method, $K_f^S$ is positive for crack-opening stresses; if $K_f^S$ is negative or zero, $\rho$ and $V$ are set to zero and 1 respectively, with no further steps necessary. If secondary stresses act alone then $\rho$ and $V$ are set to zero and 1 respectively, and $K_f^S$ is set to equal $K_f^S$ with no further steps taken (British Energy Generation Limited 2000).

Analogous to the simplified method, the ratio $K_f^S/(K_f^p/L_r)$ is then evaluated. A new parameter, $\phi$, is introduced, derived from the Option 1 FAC. The relationship between $\phi$ and the applied primary stress is shown in Figure 2-40.
Values of $\phi$ and $\psi$ are provided in look-up tables. To evaluate $\rho$, the following is performed:

- The parameter $\psi$ should be looked up in supplied tables of values, using the parameters $K_j^S / (K_i^P / L_r)$ and $L_r$.
- If $K_j^k = K_i^k$ then $\rho$ is set equal to $\psi$. If $K_j^k < K_i^k$, $\rho$ can be set equal to $\psi$ to obtain a conservative result.
- The parameter $\phi$ is obtained from supplied look-up tables, using the parameters $K_j^S / (K_i^P / L_r)$ and $L_r$.
- The $\rho$-parameter is evaluated using the relation in Equation 2-42.

$$\rho = \psi - \phi \left( \frac{K_j^S}{K_i^S} - 1 \right) \quad 2-42$$

Alternatively, to evaluate $V$, the following steps should be taken:

- The parameter $\xi$ should be looked up in supplied tables of values, using the parameters $K_j^S / (K_i^P / L_r)$ and $L_r$.
- $V$ is evaluated using Equation 2-43.

$$V = \frac{K_j^S}{K_i^S} \xi \quad 2-43$$
The effective stress intensity factor due to secondary loading alone, \( K_J^S \), can be evaluated using several approaches. The R6 procedure explains that \( K_J^S \) is related to the \( J \)-integral for secondary loading, \( J^S \), through Equation 2-44.

\[
K_J^S = \sqrt{E'J^S} \tag{2-44}
\]

Where \( E' = E \) under plane stress conditions and \( E' = \frac{E}{1-\nu^2} \) under plane strain conditions (British Energy Generation Limited 2000).

\( K_J^S \) can be evaluated through cracked-body analysis, using FEA, to calculate \( J^S \) and thus \( K_J^S \). Alternatively, an approximation to \( K_J^S \) can be evaluated through uncracked-body analysis, which requires an effective crack size to be calculated. In situations where secondary stresses are low and elastic follow-up is considered insignificant, \( K_J^S \) can be used instead of \( K_J^S \), reducing the conservatism of the method when small primary stresses are present (British Energy Generation Limited 2000).

The SIF due to combined loading is accounted for on the FAD, using Equation 2-45 and Equation 2-46.

\[
K_r^p = \frac{K_i^p}{K_{mat}} \tag{2-45}
\]

\[
K_r^S = \frac{K_i^S}{K_{mat}} + \rho \tag{2-46}
\]

The assessment point is defined using \( L_r \) and \( K_r = K_r^p + K_r^S \). Alternatively, if the \( V \) parameter is used, \( K_r \) is defined through Equation 2-47.

\[
K_r = \frac{K_i^p + VK_i^S}{K_{mat}} \tag{2-47}
\]
2.7.4.6 Alternative approaches

The EPRI-GE $J$-estimation scheme (Kumar et al. 1981) provides a compendium of $J$-integral solutions for common component geometries and defects; the results of an extensive programme of work calculating the CDF in the EPFM regime using FEA. The procedure incorporates the effect of secondary stresses on $J$, providing CDF solutions for combined loading, alleviating the need for an assessor to consider plasticity interaction factors. It is noted in R6, however, that the EPRI procedure is limited to the geometries detailed in the document (British Energy Generation Limited 2000).

2.7.4.7 The $g$-parameter

The $g$-parameter has been developed by James et al. (2009), to reduce the conservatism of the $\rho$-parameter for combined loading in ductile materials, such as austenitic stainless steels. It is based on the Option 2 FAC, requiring the stress-strain data of the material under assessment.

James et al. (2009) show that the elastic-plastic crack driving force, $K_J$, in an R6 assessment can be expressed as Equation 2-48.

$$K_J = \frac{K_J^P + K_J^S}{f(L_r) - \rho}$$  \hspace{1cm} 2-48

James et al. (2009) developed Finite Element models of a cracked cylinder containing both weld residual stress and thermal bending stresses. Defining $K_J^P = \frac{K_I}{f(L_r)}$, Equation 2-48 was rearranged to produce Equation 2-49.

$$g = \frac{K_I - K_J^P}{K_I^S}$$  \hspace{1cm} 2-49

Equation 2-49 is the ratio of the difference between the total elastic-plastic CDF and the CDF due to primary loading, and the elastic CDF due to secondary loading. $g$ as a function of $L_r$ was plotted for each FE model, comprising a range of crack sizes and geometries (James et al. 2009), an example of which is shown in Figure 2-41.
The Option 2 FAC, Equation 2-30, can be expressed in terms of the reference stress, Equation 2-50.

\[ f_2(L_r) = \left[ \frac{E \varepsilon_{ref}}{\sigma_{ref}} + \frac{L_r^2 \sigma_{ref}}{2E \varepsilon_{ref}} \right]^{-1/2} \]

A curve fit was applied to plots of \( g \) vs. \( L_r \), similar to Figure 2-41, for different crack depths, residual stress fields and thermal bending stresses, from which a definition of \( g \) in terms of the reference stress was obtained, Equation 2-51.

\[ g = \left[ \frac{E \varepsilon_{ref}^{mod}}{\sigma_{ref}^{mod}} + A \left( \frac{\varepsilon_{ref}^{mod}}{\sigma_{y}} \right)^2 \right]^{-1/2} \]
The modified reference stress, $\sigma_{ref}^{mod} = \frac{\sigma_{ref}^p}{B}$ is dependent on the primary reference stress and a constant, $B$. James et al. (2009) defined $B$ as the value of $L_r$ at the skeletal point in Figure 2-41, which provided the most conservative value. The constant $A$ is defined by Equation 2-52, the ratio of in-plane stress, $\sigma_{in\text{-}plane}$ to the von Mises stress, $\sigma_{VM}$.

$$A = -\frac{\sigma_{in\text{-}plane}}{\sigma_{VM}} \quad 2-52$$

For the range of cracks and geometries modelled, James et al. (2009) chose the lowest-bound values of $A$ and $B$, as a conservative approach: $A = 0.8$ and $B = 1.25$. Applying the $g$-parameter to the R6 simplified method reduced the conservatism of $K_r$ evaluated for a defect in a weld residual stress field in a model of a 316L stainless steel pipe. The R6 simplified method over-predicted $K_r$ by 50%; the approach using the $g$-parameter reduced this to an over-prediction of 10%, Figure 2-42.

For consistency with the R6 $V$-factor approach, the $g$-parameter can be expressed as the term $V_g$, defined in Equation 2-53.

$$V_g = g \frac{K_i^S}{K_j^S} f(L_r) \quad 2-53$$
Figure 2-42 - Comparison of R6 Option 3 and $g$-parameter approach on calculating the CDF for a circumferentially cracked pipe under a thermally induced bending stress (James et al. 2009).

2.7.5 Summary

The Failure Assessment Diagram, a powerful graphical means of calculating the proximity of a component to failure, was introduced. The use of different levels of assessment, ranging from Option 1 to Option 3, was discussed. The way R6 accounts for residual stress was described, followed by a description of the use of interaction parameters to account for the relationship between primary and secondary loads. Finally, the recently developed $g$-parameter was introduced as a development of the current interaction parameter used in R6.
2.8 Test programmes and recent work

This section reviews recent work investigating the effects of residual stress on ductile fracture and the implications for structural integrity assessments. This is followed by a review of work investigating the residual stress field and tensile and fracture toughness properties of narrow-gap pipe welds. A review of work investigating the residual stress retained in fracture mechanics specimens extracted from welded components can be found at the end of this section.

2.8.1 Effect of residual stress on structural integrity assessments

The effect of residual stress on ductile fracture has been investigated in a number of test programmes. For example, in Figure 2-43, tests were performed on aluminium plates containing different levels of residual stress (Ainsworth et al. 2000). Although it is understood that residual stress can cause failure at lower loads than would be the case if such stresses were not present, the extent to which residual stresses reduce the ductile fracture toughness of a material is not fully understood.

![Figure 2-43 - Load-carrying capacities of aluminium alloy plates with and without residual stress present (Ainsworth et al. 2000).](image)

Residual stress fields created by welds, in particular, have caused problems. Sharples and Gardner (1996) calculated instability loads for through-thickness
cracks in welded 316-stainless steel plates using the R6 procedure, yielding conservative under-predictions of instability load when compared to experimental data. However, even when the residual stress across the weldment had not been incorporated into the calculations, the conservatism remained.

The problem of accurately predicting ductile failure in girth welds has been raised by other groups. The European Network for Evaluating Structural Components (NESC) conducted a research programme, the NESC-III Project, investigating the issues surrounding structural integrity assessments of Dissimilar Metal Welds (DMWs). The project brought the work of two projects, the Assessment of Aged Piping Dissimilar Metal Weld Integrity (ADIMEW) project and the Structural Integrity of Bi-Metallic Components (BIMET) together (Taylor et al. 2006).

The ADIMEW project performed tests on a dissimilar metal girth-welded pipe of 316L stainless steel and A508 ferritic steel, shown in Figure 2-44. The programme involved a full-scale test, by applying a bending moment to initiate crack growth in the weldment, Figure 2-44.

![Dissimilar metal weld cross-section and test arrangement used in the ADIMEW project](image)

Figure 2-44 - Dissimilar metal weld cross-section and test arrangement used in the ADIMEW project (Taylor et al. 2006).

As part of the test programme, fracture mechanics specimens were extracted from the DMW and tested in several European test houses. A range of specimens, including C(T) and SE(B) specimens, were used to obtain J-R curves for different locations in the DMW, Figure 2-45. A large variation in fracture toughness was measured, with $J_{0.2BL}$ values ranging from 59 kJ/m$^2$ to 100 kJ/m$^2$. 
The weld residual stress was measured using a range of techniques including surface hole drilling and neutron diffraction, shown in Figure 2-46.
The residual stress data were incorporated into an R6 assessment to estimate the bending moment required for crack initiation in the DMW. The assessment performed Option 1, 2 and 3 analyses both with and without the measured residual stress present. The assessment used the fracture toughness of the DMW from two different fracture toughness tests, which measured $J_{\text{mat}}$ equal to 67 kJ/m$^2$ and 300 kJ/m$^2$. The results are shown in Figure 2-47.

![Figure 2-47 - Estimates of critical crack initiation moment of the ADIMEW DMW. The assessments included all three levels of R6 analysis and considered crack initiation with and without residual stress present (Taylor et al. 2006).](image)

It can be seen from Figure 2-47 that the R6 approach can be highly conservative; for a fracture toughness of 67 kJ/m$^2$, the R6 Option 1 assessment predicted a critical moment of 900 kNm, compared to the measured initiation moment of 2,000 kNm, a difference of 76%. The assessments performed without incorporating residual stress proved less conservative; an Option 3 assessment of the initiation moment for a material with a fracture toughness of 300 kJ/m$^2$ agreed with the measured initiation moment to within 11%.

The ADIMEW project has highlighted several challenges when characterising the properties of welded components. Obtaining representative measurements of fracture toughness is not always straightforward; $J$-$R$ curves obtained from test specimens extracted from the same component can show great variation, leading to a high level of uncertainty of the representative measure of the fracture toughness.
of a weld. Incorporating such data into current assessment procedures, such as R6, can therefore lead to highly conservative results, particularly when considering the effects of weld residual stress.

### 2.8.2 Characterisation of narrow-gap weld residual stress fields

Liu et al. (2011), Jang et al. (2010) and Yang et al. (2011) simulated the NG welding process using FEA, from which residual stress fields were modelled. Bouchard (2007) compiled residual stress data for a range of girth-butt pipe welds used in the UK nuclear industry, which included data for an NG weld. A summary of the welds investigated by these authors is shown in Table 2-1.

<table>
<thead>
<tr>
<th>Weld type</th>
<th>Parent material</th>
<th>Weld material</th>
<th>Thickness (mm)</th>
<th>No. weld passes</th>
<th>Residual stress measurement technique</th>
</tr>
</thead>
<tbody>
<tr>
<td>Liu et al. (2011)</td>
<td>NG butt weld</td>
<td>304L</td>
<td>316L</td>
<td>70</td>
<td>73</td>
</tr>
<tr>
<td>Jang et al. (2010)</td>
<td>&quot;</td>
<td>316L</td>
<td>308L</td>
<td>76</td>
<td>42</td>
</tr>
<tr>
<td>Yang et al. (2011)</td>
<td>&quot;</td>
<td>304L</td>
<td>308L</td>
<td>76</td>
<td>42</td>
</tr>
<tr>
<td>Bouchard (2007)</td>
<td>&quot;</td>
<td>316H</td>
<td>316L</td>
<td>62</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 2-1 - Summary of weldments analysed by Liu et al. (2011), Jang et al. (2010), Yang et al. (2011) and Bouchard (2007).

The NG weld geometry and welding method modelled by Jang et al. (2010) and Yang et al. (2011) is shown in Figure 2-48. Both groups simulated the welding process of an NG weld on a 76 mm thick, Korean PWR primary circuit pipe, performing an axisymmetric 2-D analysis using the Abaqus FEA program. Liu et al. (2011) modelled an NG weld of 73 passes in a pipe 70 mm thick using the ANSYS FEA program; the modelled residual stress field was complemented by measurements of the residual stress on the outer surface of the pipe using surface hole drilling.
The geometry of the weld investigated by Bouchard (2007) is shown in Figure 2-49, designated S5NG. The residual stress was measured using the DHD technique on the weld centreline and through the weld HAZ.

Figure 2-48 - NG weld cross section, modelled by Yang et al. (2011). The outer groove was welded first, followed by the inner groove.

Figure 2-49 - Cross-section of an NG pipe girth weld, designated S5NG (Bouchard 2007).
Figure 2-50 shows that the NG weld stress field assessed by Liu et al. (2011) had a peak width of approximately 40 mm (20 mm either side of the weld centreline), beyond which the hoop stresses dropped from approximately 300 MPa to under 100 MPa. Liu et al. (2011) explained that surface hole-drilling is applicable to situations where stresses do not vary greatly with depth, or have a magnitude below 50% of the yield strength of the material. As a result, a discrepancy exists between the simulated residual stress and the measured residual stress.

Figure 2-50 - Axial and hoop residual stress distribution on the outer surface of an NG pipe weld, including surface-hole drilled measurements and simulated residual stresses (Liu et al. 2011).
The outer surface residual stress simulated by Yang et al. (2011) is shown in Figure 2-51. Both axial and hoop stresses are tensile in nature, with a profile similar to the simulations of Liu et al. (2011). The width of the peak hoop and axial stresses are approximately 80 mm, greater than the peak simulated by Liu et al. (2011). The magnitude of the stresses also differs considerably: Jang et al. (2010) show a peak hoop stress of 260 MPa, Liu et al. (2011) show a peak hoop stress of almost 600 MPa.

Figure 2-51 - Simulated axial and hoop residual stress field on the outer surface of an NG pipe weld (Yang et al. 2011).
The simulation of the stress field on the inner surface of the pipe by Liu et al. (2011) is shown in Figure 2-52. According to the simulation, the hoop residual stresses up to 200 mm either side of the centreline are compressive. The axial stresses are also compressive, except for a region very close to the centreline, where the stresses become highly tensile. The generally compressive nature of the residual stress on the inner surface is desirable, as the compressive stresses aid crack closure and lower the susceptibility of inner surface defects to SCC.

![Graph showing simulated axial and hoop residual stresses on the inner surface of an NG pipe weld](image)

Figure 2-52 - Simulated axial and hoop residual stresses on the inner surface of an NG pipe weld (Liu et al. 2011).
Figure 2-53 shows the residual stress on the inner surface of the pipe as simulated by Yang et al. (2011). The solid lines indicate a simulation of the technique used in practice to weld the pipes, welding the outer diameter of the pipe before welding the inner diameter. Unlike the results of Liu et al. (2011), Yang et al. (2011) show that the axial residual stresses are tensile on the inner surface of the pipe. The width between the two local maxima is approximately the same, at 40 mm either side of the weld centreline. The hoop stresses simulated by Yang et al. (2011) are also different to those of Liu et al. (2011); according to Yang et al. (2011), the hoop stresses are tensile, 50 mm either side of the weld centreline, whereas the hoop stresses in the model by Liu et al. (2011) are mostly compressive.

Figure 2-53 - Simulated axial and hoop residual stress field on the inner surface of an NG pipe weld (Yang et al. 2011).
Figure 2-54 shows the simulated through-wall residual stress distribution by Liu et al. (2011). Axial residual stresses are compressive on the inner surface, becoming tensile 60% through the thickness of the pipe. Peak axial stresses are +250 MPa and -200 MPa. The hoop stresses are measured to be slightly compressive at the inner surface, but beyond 5% of the thickness become tensile in nature, peaking at 580 MPa, close to the yield strength of the weld material.

Figure 2-54 - Through-wall simulated residual stress distribution on the weld centreline for an NG pipe weld (Liu et al. 2011).
The through-wall residual stress distributions of Yang et al. (2011) are shown in Figure 2-55. The axial residual stresses are identical to that of Jang et al. (2010), as part of the same programme of work. The hoop residual stresses are tensile throughout the wall thickness, largely in agreement with Liu et al. (2011), although peak tensile stresses vary: Yang et al. (2011) simulated peak tensile hoop stresses at two maxima, 480 MPa and 350 MPa respectively. Liu et al. (2011) show a single tensile hoop stress peak, close to the OD, of approximately 580 MPa.

Figure 2-55 - Simulated axial and hoop residual stress field through the wall thickness of an NG pipe weld, on the weld centreline (Yang et al. 2011).
The residual stress profile for the NG weld, S5NG, is shown in Figure 2-56 (Bouchard 2007). The through-wall distributions are sinusoidal in nature. The hoop stress is compressive on the inner surface and tensile on the outer surface, with maxima of -150 MPa and +280 MPa. The axial residual stress is tensile at the inner surface, becoming compressive between 10% and 70% of the wall thickness, and returns to being tensile at the outer surface, with maxima of approximately -150 MPa and +250 MPa.

Figure 2-56 - Axial and hoop residual stress field through the wall thickness of an NG (S5NG) pipe girth weld (Bouchard 2007).
The residual stress profiles of Bouchard (2007) differ somewhat from the simulated through-wall profiles of Liu et al. (2011), Jang et al. (2010) and Yang et al. (2011), particularly where hoop stresses are concerned. The profiles measured by Bouchard, which are sinusoidal in nature, range from compressive at the inner surface to tensile at the outer surface. This contrasts with the hoop stresses that were shown to be tensile throughout the wall thickness in the numerical models.

### 2.8.3 Narrow gap weld mechanical properties

Due to the relatively recent development of narrow-gap welding technology in the nuclear industry, there is relatively little published work investigating the mechanical properties of NG welds. Jang et al. (2010) note that despite recognition of the advantages of NG weld technology, the mechanical properties of NG welds have yet to be fully characterised. Jang et al. (2010), Yang et al. (2011) and Ould et al. (2009) have investigated the tensile and fracture toughness properties of NG weld materials.

#### 2.8.3.1 Tensile properties

Jang et al. (2010) extracted round tensile specimens from a sample of 308L stainless steel NG weld material and neighbouring 316L parent material. The yield strength and tensile strength of each specimen is shown in Figure 2-57. Across the weldment the tensile strength shows relatively little variation. The results show that the work-hardening effect of welding has generally raised the yield strength of the weld material in the range of 320 MPa to 350 MPa, compared to a range in yield strength for the parent material of 210 MPa to 360 MPa.
Yang et al. (2011) extracted tensile specimens in a similar fashion, testing specimens at 177°C and 316°C (respectively, the hot standby temperature and the operating temperature of a Korean PWR). Ould et al. (2009) performed tensile tests on a 304L steel NG pipe weld, with thickness 70 mm and weld material of 316L steel. The results of the tensile tests from Yang et al. (2011) and Ould et al. (2009) are summarised in Table 2-2.

<table>
<thead>
<tr>
<th>Material</th>
<th>Ould et al. (2009)</th>
<th>Yang et al. (2011)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Test Temperature (°C)</td>
<td>~20</td>
<td>177, 316</td>
</tr>
<tr>
<td>Yield strength (MPa)</td>
<td>260, 479</td>
<td>391, 357</td>
</tr>
<tr>
<td>Tensile strength (MPa)</td>
<td>528, 621</td>
<td>461, 423</td>
</tr>
<tr>
<td>Elongation (%)</td>
<td>65.5, 39.5</td>
<td>31.7, 27.3</td>
</tr>
<tr>
<td>Reduction of Area (%)</td>
<td>79.0, 71.5</td>
<td>-</td>
</tr>
</tbody>
</table>

Table 2-2 - Tensile test data NG pipe welds by Ould et al. (2009) and Yang et al. (2011).

The tensile results do not show much consistency in yield strength or tensile strength; this may be accounted for by the different test temperatures, the grade of steel tested and the test method. However, the results of Jang et al. (2010) are reasonably consistent with Yang et al. (2011); the yield and tensile strengths agree to within 10%.
2.8.3.2 Fracture toughness properties

Yang et al. (2011) extracted C(T) specimens from weld test coupons to measure the fracture toughness of the 308L NG weld material. A diagram of specimen extraction is shown in Figure 2-58. Fracture toughness tests were performed using the unloading compliance technique; the fracture toughness parameter $J_{0.2BL}$ was chosen to characterise the weld material fracture toughness.

Figure 2-58 - Orientation of the C(T) and R(T) specimens extracted from NG weld test coupons (Yang et al. 2011).

Ould et al. (2009) measured the fracture toughness of an NG weld using C(T) specimens, extracted from the weld material in two different orientations, illustrated in Figure 2-59. The ASTM E1820 standard was followed; the multi-specimen technique was used to test the specimens at 300°C, obtaining values for $J_{0.2}$, $J_{0.2BL}$ and $J_{1.0}$. 

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Figure 2-59 - Orientation of C(T) specimens extracted from a 316L NG weld (Ould et al. 2009).

J-resistance curves from the two test programmes are shown in Figure 2-60 and Figure 2-61. A summary of the fracture toughness measurements is provided in Table 2-3.

![J-resistance curves](image)

Figure 2-60 - J-resistance curves for 308L NG weld fracture toughness tests (Yang et al. 2011).

<table>
<thead>
<tr>
<th>Material</th>
<th>Jang et al. (2011)</th>
<th>Ould et al. (2009)</th>
</tr>
</thead>
<tbody>
<tr>
<td>308L</td>
<td></td>
<td>316L</td>
</tr>
<tr>
<td>Test temperature (°C)</td>
<td>177</td>
<td>316</td>
</tr>
<tr>
<td>Average J_{0.2BL} (kJ/m²)</td>
<td>1307</td>
<td>934</td>
</tr>
</tbody>
</table>

Table 2-3 - Summary of fracture toughness measurements made by Ould et al. (2009) and Yang et al. (2011).
The fracture toughness values obtained from the two test programmes varies greatly, which may be due to the test method, specimen geometries and material composition. The $J$-resistance curve of Ould et al. (2009) is fitted to data which exhibits a lot of scatter. There is also insufficient data to establish whether the specimen orientation had an effect on measured fracture toughness.

Despite using 25 mm thick specimens, Ould et al. (2009) noted that the values of $J$ obtained from the multi-specimen tests are not valid according to the ASTM E1820 standard; both the $J_{\text{max}}$ and $\Delta a_{\text{max}}$ parameters were exceeded, due to the high toughness of the material relative to the specimen size – widespread plasticity prevented a sufficiently conservative measure of fracture toughness from being measured.

### 2.8.4 Retained residual stress in test specimens

To measure the fracture toughness of welds, specimens, such as C(T) and SE(B) test pieces, are extracted from full-sized components. The level of residual stress that is retained in a test piece is often unknown. The size and geometry of the original residual stress field, the full-sized component, the test piece and the mechanical properties of the material can influence the level of retained residual stress. High
levels of retained residual stress can influence the measured fracture toughness of a material. For example, a material exhibiting low work-hardening behaviour could retain a high level of residual stress up to fracture, as there is little plasticity available to redistribute residual stresses.

Altenkirch et al. (2009) studied the relaxation of weld residual stresses when test-pieces were extracted from welded plates. Two plates of aluminium alloy AA2098 were joined using the Friction Stir-Welding (FSW) technique. The process was repeated for aluminium alloy AA7449 plates. Two S355 steel plates were joined: one using a hybrid-laser Metal Inert Gas (MIG) weld, the other, a pulsed MIG weld. Different plate dimensions and weld parameters were used to create a range of residual stress fields. The residual stress fields were measured using high energy XRD. The plates were then sectioned, some longitudinally, some transversely, to reduce the plate size, and the residual stress profile was re-measured. The plates continued to be reduced in size, to determine the level of relaxation of weld residual stress upon sectioning.

On sectioning the plates longitudinally (reducing the length of the plate), Altenkirch et al. (2009) noted that the original transverse stresses remained largely unaffected. The longitudinal stresses did reduce in magnitude, although the width of the tensile
peak toward the weld centreline remained relatively constant, shown in Figure 2-63.

![Figure 2-63](image)

Figure 2-63 - Longitudinal residual stress distributions in the four welded plates (Altenkirch et al. 2009).

By normalising the plate length and normalising the residual stress distribution, Altenkirch et al. (2009) showed that the longitudinal weld residual stress was unaffected by material removal up to a point, beyond which the stress field decayed exponentially.

![Figure 2-64](image)

Figure 2-64 - Normalised weld residual stress versus normalised plate length (Altenkirch et al. 2009).
Altenkirch et al. (2009) developed an empirical relation through observation of the change in the residual stress fields on sectioning. By relating the width of the tensile peak, $w_t$, to the residual stress in a (theoretically) infinite residual stress field, $\sigma_0$, the longitudinal residual stress, $\sigma_{long}$, could be obtained for a weld of length $l$, Equation 2-54, plotted in Figure 2-65.

$$\sigma = \sigma_0 \left[ 1 - \exp \left( \frac{1}{3} \left( -\frac{l}{w_t} \right) \right) \right]$$  

Figure 2-65 – Normalised longitudinal residual stress versus normalised plate length based on the relation described by Equation 2-54 (Altenkirch et al. 2009).

On transverse sectioning of the plate, Altenkirch et al. (2009) noted that transverse residual stresses were relatively unaffected by narrowing of the plates. The longitudinal residual stresses decreased, redistributing to maintain self-balance within the plate. This manifested itself as a downward shift in plots of the longitudinal stress field, although the shape of the field was maintained, as shown in Figure 2-66.
Altenkirch et al. (2009) developed a relation to evaluate the retained longitudinal residual stress on transverse sectioning, based on the width of the tensile stress peak, $w_t$, and the width of the plate, $w$, Equation 2-55.

$$\sigma = \sigma_0 \left[1 - \frac{w_t}{w}\right] \quad 2-55$$

Altenkirch et al. (2009) acknowledged that the observations applied only to thin plates. In thicker plates, where triaxial stress states exist, the equations used to describe the change in residual stress on specimen extraction may not hold true; Altenkirch et al. noted that further research is required to see if Equation 2-54 and Equation 2-55 apply for different component geometries.

Hurlston et al. (2012) applied a stress partitioning technique outlined in the R6 procedure (British Energy Generation Limited 2000) to calculate the retained transverse residual stress in SE(B) specimens extracted from a welded A508 steel plate. The technique separated the residual stress field into membrane, bending and self-equilibrating components, $\sigma_m$, $\sigma_b$, and $\sigma_{eq}$, respectively, defined in Equations 2-56, 2-57 and 2-58.

$$\sigma_m = \frac{1}{W} \int_{z=0}^{z=W} \sigma(x) \, dx \quad 2-56$$
Where \( W \) is the specimen width and \( x \) is the distance through the wall thickness of the pipe. Hurlston \textit{et al.} (2012) compared predictions of retained transverse residual stress in specimens using the stress partitioning technique to FEA simulations of specimen extraction from the welded plate, Figure 2-67. The results show good agreement between the retained residual stress calculated through the stress partitioning approach and the numerically simulated results.

\[
\sigma_b = \frac{6}{W^2} \int_{z=0}^{z=W} \sigma(x) \left( \frac{W}{2} - x \right) dx
\]

\[
\sigma_{eq}(x) = \sigma(x) - \sigma_m - \sigma_b(x)
\]

Figure 2-67 - Comparison of retained transverse residual stress in bend specimens, calculated using the stress partitioning technique with FE simulation of specimen extraction from an A508 steel welded plate (Hurlston \textit{et al.} 2012).
2.8.5 Summary

There is relatively little published data focusing on the residual stress field and tensile and fracture toughness properties of NG welds. Most NG weld residual stress measurements have been made through numerical modelling approaches; there is very limited information on NG weld residual stress fields measured experimentally. The mechanical properties of a few NG welds have been measured, but there is little like-for-like data from which comparisons may be drawn. In particular, the fracture toughness of NG welds has not been fully established, with little consistency between the results of the different groups that have published fracture toughness data. Finally, the level of weld residual stress relaxation on extraction of test specimens has been investigated for thin welded plates. However, this work is limited to friction stir-welded plates; no work has yet been undertaken to establish the relationship between specimen size, orientation and retained residual stress for different types of welded component.
2.9 Knowledge gaps

The literature review has shown that there are well developed methods to characterise the residual stress state in as-welded components, and test methods to derive mechanical and fracture toughness properties of welds. Defect assessment methods have been developed to quantify the proximity to failure of a welded structure containing a defect. However, the application of these approaches to the characterisation and assessment of narrow gap welds is sparse. In particular:

- Austenitic narrow-gap pipe weld residual stress fields have yet to be fully measured. There is very little published data on the size and range of NG weld residual stress fields obtained through experimental work,
- There is little published data on the mechanical properties of austenitic NG welds, in particular the weld fracture toughness,
- The effect of test specimen orientation on the measured fracture toughness of NG welds is unknown,
- The level of retained residual stress in test specimens is not fully understood, nor is its effect on the measured fracture toughness of a material,
- The g-parameter has yet to be applied to a model of a real pipe-girth weld containing an experimentally measured residual stress field.

The work undertaken for this PhD project has sought to address these issues. The aim of the research is to provide a better understanding of the properties of austenitic stainless steel narrow-gap girth welds and to suggest a suitable approach to obtain conservative measurements of the weld properties for use in structural integrity assessments. As described in the following chapters, these knowledge gaps have been addressed through a combination of experimental and modelling approaches. The original weld residual stress state of as-received NG pipe welds are characterised using DHD and neutron diffraction. Mechanical properties are characterised using tensile and fracture mechanics tests, and the level of retained residual stress in test specimens is characterised using DHD. Finally, the assessment of residual stresses on the structural behaviour of a narrow-gap pipe weld is predicted using FEA.
3 Experimental methodology

3.1 Supplied sections of pipe weld

3.1.1 Material and dimensions

Two sections of narrow-gap girth-welded pipe were supplied. The parent material was made of 304L stainless steel and the weld material of 308L stainless steel. One of the pipes – the long section – was 200 mm in length. The second pipe – the short section – was 70 mm in length. The dimensions of the two pipe sections are summarised in Table 3-1.

<table>
<thead>
<tr>
<th></th>
<th>Long Pipe</th>
<th>Short Pipe</th>
</tr>
</thead>
<tbody>
<tr>
<td>Outer diameter (mm)</td>
<td>358</td>
<td>358</td>
</tr>
<tr>
<td>Thickness (mm)</td>
<td>33</td>
<td>33</td>
</tr>
<tr>
<td>Approx. length (mm)</td>
<td>200</td>
<td>70</td>
</tr>
</tbody>
</table>

Table 3-1 - Summary of pipe dimensions.

The geometries of the pipe are shown in Figure 3-1 and Figure 3-2. Note the girth weld, positioned halfway along the length of each pipe.
3.1.2 Chemical composition testing

Samples of parent and weld material were extracted from the short pipe section. The material underwent chemical composition analysis at Intertek Ltd, Derby, using the Inductively-Coupled Plasma Optical Emission Spectroscopy (ICP-OES) technique. The measured weight percent of carbon, manganese, silicon, sulphur, phosphorous, copper, nickel, chromium, molybdenum, niobium, vanadium, aluminium and titanium were measured and compared to the 304L and 308L specifications in the ASTM A312/A312M specification (American Society for Testing and Materials 2011a).

3.1.3 Weld configuration

Both pipes had identical weld configurations, shown schematically in Figure 3-3. The cold wire fully orbital Tungsten Inert Gas (TIG) weld comprised approximately 20 passes in the form of full-width weaves. The weld consumable for both the root insert and fill was 308L stainless steel. Due to issues of confidentiality, the exact welding parameters were not made available.
3.2 Metallography

A slice of material was removed from the short pipe section, containing both parent and weld material, Figure 3-4. Six smaller sections were extracted from the removed slice of material, Figure 3-5. Three sections were mounted in resin so that the parent material, weld material and the weld-parent boundary could be examined in the Radial-Axial (R-A) plane. The other three sections were mounted so that the parent, weld and parent-weld material could be examined in the Radial-Hoop (R-H) plane. These planes were chosen because their orientation would elucidate the microstructure through which cracks would grow in fracture mechanics specimens extracted from the pipes.

Figure 3-4 - Slice of material extracted from the short pipe section.
The surface of each specimen was ground up to P4000 grade (surface finish with a median value of 2.5 µm). Each specimen was then polished to ¼ µm finish, and examined under an optical microscope to check for the absence of polishing defects.

To reveal the grain structure of the metallographic samples, the specimens were etched using 10% concentration oxalic acid. A current of 2A at a potential of 6V was applied to each specimen for 30 seconds. Each specimen was then washed in distilled water and re-examined under the microscope and re-etched according to the visibility of the grain structure. The parent material specimens were etched for a total of 90 seconds, the weld material specimens etched for 60 seconds. For the specimens covering the weld-parent boundary, etching lasted for 60 seconds to reveal the weld microstructure. To reveal the parent microstructure more clearly, packaging tape was used to mask the weld material, preventing further etching from taking place. The specimens were etched for a further 30 seconds to reveal the parent microstructure.

To obtain the final micrographs an Olympus BX15M microscope was used to examine the specimens. The ASTM E112 (American Society for Testing and Materials 2010) procedure was used to determine the grain size of the material using the line-intercept method. Five lines of known length (chosen to be 300 µm) were drawn onto each micrograph, the location and orientation of which were chosen at random. The number of grain boundaries intercepted by each line was counted (triple grain boundary junctions were counted as 1.5). Dividing the sum of the intercepts across all lines by their total length yielded the average grain size for
each micrograph. The results from each micrograph were averaged together to produce an overall average grain size for the material. ASTM grain-size charts were also used to establish the grain size of the material and compared with the line-intercept approach.

### 3.3 Residual stress measurements

Two techniques were used to characterise the residual stress field in the long section of pipe. Deep Hole Drilling (DHD), performed at the Mechanical Engineering department at the University of Bristol, was used to measure the axial and hoop weld residual stresses through the wall thickness of the pipe, on the weld centreline. Neutron Diffraction (ND) was undertaken using the ENGIN-X instrument on the ISIS beamline at the Rutherford-Appleton Laboratory, Harwell, Oxfordshire. ND was used to characterise the residual stress field across the weldment, along the axis of the pipe, and through the wall thickness of the pipe, from which comparisons with the DHD technique were drawn.

#### 3.3.1 Deep Hole Drilling

The DHD technique was applied in the following stages, outlined in Figure 3-6:

a) Reference bushes were attached to the outer and inner surfaces of the pipe using araldite. A 1.5 mm diameter reference hole was then gun-drilled on the weld centreline, through the thickness of the pipe wall, using the reference bushes as a guide

b) An air-probe travelled down the reference hole, measuring the profile of the hole in 0.2 mm increments of depth

c) A cylinder of material surrounding the reference hole was trepanned using Electro-Discharge Machining (EDM), causing the residual stresses surrounding the reference hole to relax

d) The air-probe was sent down the reference hole once again to re-measure the profile of the reference hole. Any change in diameter of the hole was due to relaxation of the trepanned material, from which the original weld residual stress was calculated.
The air probe delivered a jet of air at 0° and 180° to its axis, allowing the probe to float in the centre of the reference hole. The pressure required to centre the probe was related to the diameter of the hole. Calibration discs, containing holes of various diameters, were used to calibrate the air probe before measurements were made. The air probe was recalibrated every 3 to 4 hours, as the device was sensitive to fluctuations in the atmospheric conditions in the laboratory.

Figure 3-6 - Stages of the Deep Hole Drilling technique.

Figure 3-7 - Air probe rotational measurement positions.
At each increment of depth the air probe measured the diameter of the reference hole at 16 discrete angles, to account for the different levels of relaxation in different planes, Figure 3-7. In the case of the pipe weld, residual strains were measured in the axial and hoop directions. Care was taken not to locate the reference hole at or near the weld start/stop position, where the measured residual stresses were likely to be unrepresentative of the residual stress levels in the rest of the component.

![Figure 3-7](image)

**Figure 3-8 - Long pipe being gundrilled.**

### 3.3.1.1 DHD analysis

The reference hole diameter, $\phi_0$, was recorded as a function of the angular orientation of the air probe, $\theta$. The reference hole diameter after trepanning was recorded as the variable $\phi$. The change in reference hole diameter, $\Delta\phi$, due to trepanning is expressed in Equation 3-1.

$$\Delta\phi = \phi - \phi_0$$  \hspace{1cm} 3-1
The change in reference hole diameter was related to the residual stress field through Equation 3-2 (Mahmoudi et al. 2009).

\[ u_{rr}(\theta) = \frac{\Delta \phi}{\phi} = -\frac{1}{E} \left[ \sigma_{xx}(1 + 2\cos2\theta) + \sigma_{yy}(1 - 2\cos2\theta) + 4\sigma_{xy}\sin2\theta \right] \quad 3-2 \]

Where \( u_{rr}(\theta) \) is the normalised radial displacement, \( E \) is Young’s modulus, \( \sigma_{xx} \) is the stress component in the x-direction, \( \sigma_{yy} \) is the stress component in the y-direction and \( \sigma_{xy} \) is the shear stress.

Alternatively, Equation 3-2 can be expressed as Equation 3-3.

\[ u_{rr}(\theta) = -\frac{1}{E} \left[ \sigma_{xx}f(\theta) + \sigma_{yy}g(\theta) + \sigma_{xy}h(\theta) \right] \quad 3-3 \]

Where:

\[ f(\theta) = (1 + 2\cos2\theta) \]
\[ g(\theta) = (1 - 2\cos2\theta) \]
\[ h(\theta) = (4\sin2\theta) \quad 3-4 \]

For each increment of depth, the reference hole diameter was measured in \( n \) angular locations (in this experiment, \( n = 8 \)). Goudar et al. (2008), show that Equation 3-4 can be expressed as a matrix, Equation 3-5.

\[ u_{rr} = -\frac{1}{E} M \sigma \quad 3-5 \]

Where

\[ u_{rr} = \begin{bmatrix} u_{rr}(\theta_1) \\ u_{rr}(\theta_2) \\ \vdots \\ u_{rr}(\theta_n) \end{bmatrix}, \quad M = \begin{bmatrix} f(\theta_1) & g(\theta_1) & h(\theta_1) \\ f(\theta_2) & g(\theta_2) & h(\theta_2) \\ \vdots & \vdots & \vdots \\ f(\theta_n) & g(\theta_n) & h(\theta_n) \end{bmatrix}, \quad \sigma = \begin{bmatrix} \sigma_{xx} \\ \sigma_{yy} \\ \sigma_{xy} \end{bmatrix} \quad 3-6 \]
A least-squares fit to the strains was used to calculate the residual stresses, $\sigma$, using Equation 3-7.

$$\sigma = -EM^*u_{rr}$$  

Where:

$$M^* = (M^TM^{-1})M^T$$  

$M^*$ is the pseudo-inverse matrix of $M$ and $M^T$ is the transpose of $M$.

### 3.3.1.2 Measurement errors

Error analysis of the DHD measurements was calculated using the following method, derived by Goudar et al. (2008). The main source of error is related to the calibration of the air probe, and the resolution of the pressure gauge from which hole distortion measurements are calculated. Goudar et al. (2008) applied the propagation of errors method to quantify the uncertainty in the DHD strain measurements as follows.

The error in the change in diameter, $\delta(\Delta \phi)$, was obtained using Equation 3-9.

$$\delta(\Delta \phi) = \sqrt{\left(\frac{\partial(\Delta \phi)}{\partial \phi}\right)^2 \delta \phi^2 + \left(\frac{\partial(\Delta \phi)}{\partial \phi_o}\right)^2 \delta \phi_o^2}$$  

Which becomes:

$$\delta(\Delta \phi) = \sqrt{(\delta \phi)^2 + (\delta \phi_o)^2}$$  

The error in $u_{rr}$ was obtained using the same approach, giving Equation 3-11.

$$\frac{\delta(u_{rr})}{u_{rr}} = \sqrt{\left(\frac{\delta(\Delta \phi)}{\delta \phi}\right)^2 + \left(\frac{\delta \phi_o}{\phi_o}\right)^2}$$  

Goudar et al. (2008) go on to show that for a particular depth, $z$, and for the number of angular measurements made, $n$, the stresses are obtained through Equation 3-12.
The error in stress is given by Equation 3-13.

\[ \delta(\sigma_{xx}) = E \sum_{i=1}^{n} \left( \frac{\partial \sigma_{xx}}{\partial u_{ni}} \right)^2 \delta u_{ni}^2 \] \hspace{1cm} 3-13

Which evaluates to:

\[ \delta(\sigma_{xx}) = E \sum_{i=1}^{n} M_{i1}^2 \delta u_{ni}^2 \] \hspace{1cm} 3-14

For the DHD measurements performed on the pipe, \( \sigma_{xx} \) is the residual stress in the hoop direction. Equations 3-12, 3-13 and 3-14 are equivalent for the residual stress in the \( \sigma_{yy} \) (axial) and \( \sigma_{xy} \) (hoop-axial shear) components of stress.
3.3.2 Neutron Diffraction

3.3.2.1 Specimen preparation

An area of the pipe weld was selected upon which ND linescans were to be performed, away from the weld start/stop zones and the DHD location. The periphery of the measurement area was marked by gluing six ball bearings to the outer surface of the pipe. The ball bearings formed fiduciary markers, which acted as reference points to show the spatial relationship between the surface of the pipe and its edges, Figure 3-9. The measurement area, including the ball bearings, were then sprayed in a light powder, to make the measurement area suitable for scanning with a CIMCore Co-ordinate Measuring Machine (CMM) laser scanner. The scanner, operated by hand, was passed over the whole pipe, recording the surface features of the outer and inner walls, Figure 3-10. This was used to generate a three-dimensional image of the component, which was imported into the software program SScanSS v.4.2.1, developed at the ISIS facility. The software used the 3-D scanned image to calculate the optimal position of the pipe in the experimental chamber, so that the path length of the neutron beam passing through the pipe wall could be minimised.

Figure 3-9 - Attaching ball bearings to act as fiduciary markers on the pipe weld.
The pipe was loaded onto the main stage in the experimental chamber using a crane. A CimCore CMM touch probe was used to locate the co-ordinates of the six Fiduciary markers on the pipe. The recorded co-ordinates were imported into SScanSS, which adjusted the position of the main stage to align the pipe to the optimal position.

3.3.2.2 Strain-free lattice parameter measurement

A diffraction comb was machined from a block of material extracted from the short pipe section by CNC Precision Ltd, Redditch. The block was made of parent material at the sides, with weld material running down its centre, Figure 3-11. The teeth of the comb were cut from the block using EDM; each tooth was made of both parent and weld material, with a cross-section of 4 mm$^3$, chosen to match the diffraction gauge volume used. A cutting plan is provided in Appendix B.
Machining the comb caused the residual stress across the weldment to relax, allowing the strain-free lattice parameter, \( d_0 \), to be determined by measuring the lattice-plane spacing of the stress-relieved material. Several \( d_0 \) measurements were taken, to obtain strain-free lattice plane spacings in the weld material (on the weld centreline) and in the parent material, shown in Figure 3-12.

![Diagram of scan locations](image)

**Figure 3-12 - Locations of \( d_0 \) scans, shown by the red crosses, on the diffraction comb.**

### 3.3.2.3 Measurement technique

In components of a simple geometry – such as a thin plate – it is possible to measure the strains inside the component in a straightforward fashion. The neutron beam can enter the specimen, diffract, exit the specimen and reach the two detectors, measuring two orthogonal components of strain simultaneously, Figure 3-13. To measure strain in all three principal directions, the component undergoes a linescan in one orientation, measuring two of the principal strains. Once complete, the specimen is rotated in the experimental chamber, Figure 3-14. The linescan is repeated so that the final principal strain component can be measured.
Figure 3-13 - The neutron beam passing through a small specimen. Both detectors can receive the diffracted neutron beam, allowing two components of strain to be measured simultaneously.

In thin specimens, the length of the path through the material along which the neutrons must travel is short. Because the neutrons encounter a small amount of material, the likelihood of the neutrons scattering off atoms in the material is low. Therefore more neutrons can reach the detectors unimpeded; the signal recorded
by the detectors has a high Signal to Noise Ratio, due to low attenuation of the neutron beam.

In thick-walled specimens, or components with a complex geometry, the neutron beam path length through the walls of the component can increase substantially. If the path length becomes too long, it is not possible to obtain a signal of sufficient strength at the detectors. For the thermal neutrons produced at the ISIS beamline, a path length of approximately 50 mm in stainless steel will attenuate the beam to undetectable levels.

The wall thickness of the pipe weld, and therefore the minimum path length through the body of the component, was 33 mm. The geometry of the pipe presented problems: to measure hoop and radial strains, the neutron beam had to pass through the measurement area (one wall thickness) and then through the opposite wall to reach the detector. For measuring axial strains, it was possible to align the pipe so that the beam only had to pass through one wall, allowing for the strains to be measured directly, Figure 3-15.

![Neutron beams diagram](image)

Figure 3-15 - Neutron paths for measuring axial residual strains. Only the south detector was used; neutrons were unable to reach the north detector, impeded by the pipe wall opposite the measured volume.
A novel approach was applied to overcome this problem. A technique similar to the $\sin^2 \psi$ method used in XRD was applied. The pipe was orientated on the stage in such a way that one diffracted path length was very short, the other long. The measured strain was a hoop-radial shear strain. By rotating the pipe through an angle, $\psi$, different levels of hoop-radial shear strain were obtained, Figure 3-16. Only the south detector, which received the neutrons travelling along the short path length, was used, Figure 3-17. The analysis of the shear strain data, to obtain resolved hoop and radial strains, is detailed in the following section.

Figure 3-16 - The $\sin^2 \psi$ method, showing how rotating the pipe through an angle, $\psi$, was used to measure hoop-radial shear strains.
A 4 mm³ gauge volume was chosen as a compromise: large enough to obtain sufficient neutron counts in a reasonable amount of time, but small enough for adequate resolution. To increase the SNR, the chopper, which controls the range of thermal neutrons to pass from the beamline to the ENGIN-X instrument, had its frequency increased from 25 Hz to 50 Hz, allowing a greater proportion of high-energy neutrons into the experimental chamber.
3.3.2.4 Linescan across weld at 10 mm depth from OD

To acquire sufficient strain data within the allocated time, a single linescan was performed across the weldment, parallel to the axis of the pipe. The linescan was positioned 10 mm below the outer surface of the pipe; the initial DHD results showed that this was in the region of highest tensile residual stress. To measure the peak tensile residual stress was considered important, because such stresses contribute to crack opening forces.

Figure 3-19 - Linescan performed across the weld, 10 mm below the OD.

Because the component contained a similar metal weld, the linescan was designed to lie asymmetrically across the weld centreline, in order to capture as much information on the residual strains present, whilst making the most efficient use of time. On one side, measurements started 15 mm away from the centreline in parent material, moving towards the centreline at 3 mm intervals. On the other side of the centreline, measurements were performed at 3 mm intervals to mirror those on the opposite side. After 9 mm, measurements were made at shorter intervals to capture strain data on the weld-parent boundary. Measurement intervals were gradually increased after 14 mm, up to 49 mm away from the weld centreline. The linescan is shown in Figure 3-19.
3.3.2.5 Linescan through wall thickness on weld centreline

A second linescan was performed on the weld centreline, through the pipe wall, to complement the DHD measurements. Four points were measured at 5, 15, 20 and 25 mm below the OD (the point at 10 mm had already been measured as part of the first linescan). The linescan is shown in Figure 3-20. The increase in path length through the material at the points 20 and 25 mm below the OD led to very high signal attenuation; it was not possible to measure the residual strains at those depths.

![Figure 3-20 - Linescan through weld material on the weld centreline.](image)

3.3.2.6 Neutron diffraction data analysis

Diffraction spectra data obtained from the detectors were used to measure the lattice parameter, \(d\). The code GSAS was used to apply a least squares fit to each diffraction spectrum, to elucidate the diffraction peaks, from which the peak shift, \(\Delta \lambda\), was obtained.

The strain component, \(\varepsilon\), for a given diffraction angle, \(\theta\), was obtained using the relation in Equation 3-15.

\[
\varepsilon = \frac{d - d_0}{d_0} = \frac{\Delta \lambda}{\lambda} - \cot \theta \cdot \Delta \theta
\]

3-15
For a pulsed neutron source, such as the ISIS source, \( \Delta \theta = 0 \), simplifying Equation 3-15 to Equation 3-16.

\[
\varepsilon = \frac{d - d_0}{d_0} = \frac{\Delta \lambda}{\lambda}
\]  

Equation 3-16 was used to determine the axial component of residual strain, \( \varepsilon_{ax} \), in the pipe weld, and the strain-free lattice parameter, \( d_0 \), from the diffraction comb.

For measurements of hoop and radial strain, plots of the measured lattice-plane spacing, \( d \), versus \( \sin^2 \psi \), were drawn. An example of a typical plot is shown in Figure 3-21. The ellipsoidal form of the plots was evidence of the presence of shear strains. A straight-line was fitted, using a least-squares method, to lie halfway between the positive and negative shear strains, using the Newton-Raphson algorithm, bisecting the ellipse. Extrapolating the line to values of \( \sin^2 \psi = 0 \) and \( \sin^2 \psi = 1 \) and looking up the corresponding value of \( d \) yielded the lattice plane spacing in the principal directions, the radial and hoop strains respectively. As explained in Section 2, obtaining strain components using the \( \sin^2 \psi \) method does not require a measurement of \( d_0 \).

![Figure 3-21 - Ellipsoidal lattice parameter measurements, revealing the presence of shear strains.](image)

For a given component of strain, the Poisson effect causes strains in the remaining two orthogonal directions. To account for this, a correction was applied to the calculated stress data using Equation 3-17, where \( v = 0.3 \) (Goudar et al. 2008).
Where \( \sigma_{hp} \), \( \sigma_{ax} \) and \( \sigma_{rad} \) are the components of hoop, axial and radial residual stress respectively and \( \varepsilon_{hp} \), \( \varepsilon_{ax} \) and \( \varepsilon_{rad} \) are the hoop, axial and radial components of residual strain respectively.

### 3.3.2.7 Measurement Errors

Uncertainty in measured lattice spacings can be attributed to several factors, such as detector efficiency, photomultiplier tube efficiency and chopper speed regulation. Due to the precision of these components at the ENGIN-X facility, measurement errors are extremely small when compared to errors in peak fitting, to the point where they were neglected.

To quantify the error due to peak fitting, the propagation method was used (Goudar et al. 2008). The error in axial strain, \( \delta \varepsilon_{ax} \), was obtained using Equation 3-18.

\[
\delta \varepsilon_{ax} = \sqrt{\left( \frac{\partial (\varepsilon_{ax})}{\partial d_{ax}} \delta d_{ax} \right)^2 + \left( \frac{\partial (\varepsilon_{ax})}{\partial d_0} \delta d_0 \right)^2} \tag{3-18}
\]

Which evaluates to Equation 3-19.

\[
\delta \varepsilon_{ax} = \frac{d_{ax}}{d_0} \sqrt{\left( \frac{\delta d_{ax}}{d_{ax}} \right)^2 + \left( \frac{\delta d_0}{d_0} \right)^2} \tag{3-19}
\]
The error in the axial stress component, $\delta \sigma_{ax}$, was obtained using Equation 3-20 (Goudar et al. 2008).

$$\delta \sigma_{ax} = \sqrt{\left(\frac{\partial \sigma_{ax}}{\partial \varepsilon_{ax}} \delta \varepsilon_{ax}\right)^2 + \left(\frac{\partial \sigma_{ax}}{\partial \varepsilon_{hp}} \delta \varepsilon_{hp}\right)^2 + \left(\frac{\partial \sigma_{ax}}{\partial \varepsilon_{rad}} \delta \varepsilon_{rad}\right)^2}$$  

Rearranging the triaxiality correction in Equation 3-17, gives Equation 3-21.

$$\sigma_{ax} = \frac{E}{(1 + v)(1 - 2v)} \left[(1 - v)\varepsilon_{ax} + v(\varepsilon_{hp} + \varepsilon_{rad})\right]$$  

Goudar et al. (2008) used Equations 3-20 and 3-21 to yield the measurement error in axial stress, Equation 3-22.

$$\delta \sigma_{ax} = \frac{E}{(1 + v)(1 - 2v)} \sqrt{(1 - v)^2(\Delta \varepsilon_{ax})^2 + v^2\left((\Delta \varepsilon_{hp})^2 + (\Delta \varepsilon_{rad})^2\right)}$$  

Equation 3-22 was applied similarly to obtain the errors in measured hoop and radial components of strain.

### 3.3.3 Blank SE(B) residual stress measurements

A novel approach was used to determine the level of residual stress retained in fracture mechanics specimens, by modifying the DHD technique, as follows:

a) A reference hole, 1.5 mm in diameter, was gundrilled on the weld centreline in the large pipe section.

b) The profile of the reference hole was measured using an air probe.

c) Instead of trepanning the reference hole, a test piece, similar to a blank SE(B) specimen, was extracted from the pipe weld, centred on the reference hole. This caused the weld residual stresses to partially relax.

d) The profile of the reference hole, now lying in the blank specimen, was re-measured.

e) The reference hole was trepanned, using wire-EDM, to fully relax the residual stresses.
f) The profile of the reference hole, now fully trepanned, was re-measured.

The body of the blank specimen had the same basic dimensions as a standard SE(B) specimen. However, to keep the original DHD reference points – the reference bushes – it was necessary to adapt the test piece design to keep the bushes attached, Figure 3-22.

![Test piece based on a blank SE(B) specimen to preserve reference hole.](image)

To establish the residual stress retained in the blank specimen, the profile of the reference hole measured in step (d) was subtracted from the hole profile measurement in step (f).

Subtracting the profile of the reference hole measured in step (b) from the final hole measurement in (f) enabled the original weld residual stress field to be calculated. This result was compared to the weld residual stresses measured using conventional DHD.

Two blank SE(B) test pieces, one in the axial orientation and one in the hoop orientation, were extracted from the pipe weld to compare the residual stress retained in specimens of different orientation. Shows a simplified diagram of the blank specimens in relation to the pipe weld. Cutting plans are provided in Appendix B.
3.4 Mechanical tests

3.4.1 Specimen extraction

Both the long and short pipes were sectioned to obtain a range of test specimens. The locations of the specimens were distributed across the pipe to avoid bias in the results to a particular weld location. Cutting plans are provided in Appendix B.

Round tensile, R(T), specimens were designed to the BS EN ISO 6892-1:2009 standard (British Standards Institution 2009) and were aligned with the hoop direction of the pipe. Five tensile specimens were extracted from parent material and five from weld material in the short pipe section. Two tensile specimens were also extracted from weld material in the long pipe section, to compare the tensile properties of the weld material in the different pipes. The specimens were 5 mm in diameter, 25 mm in gauge length and had M8 threaded ends. The weld tensile specimens were extracted on the weld centreline to ensure that the specimens consisted entirely of weld material. Parent tensile specimens were extracted as far away from the weldment as possible, ensuring the specimens were made of material unaffected by the welding process. Specimen drawings are provided in Appendix C.
Fracture mechanics specimens, in the form of SE(B) test pieces, were designed in accordance with ESIS P2-92 (European Structural Integrity Society 1992). In the short pipe section, five specimens were extracted from the parent material and five from the weld material, aligned with the hoop direction. In the long section of pipe, seven SE(B) specimens, aligned with the axial direction, were extracted from the weld material.

All fracture mechanics specimens had a width, $W$, of 20 mm, thickness, $B$, of 10 mm, span, $S$, of 80 mm and overall length of 92 mm. Each specimen had machined integral knife-edges and were side-grooved by 10% on each side, giving a nominal net thickness, $B_n$, of 8 mm. Side-grooves were machined with a $90^\circ$ included angle, following fatigue pre-cracking (Section 3.4.4.1) to ensure that closure of the side-grooves at the back face of the specimen did not occur at large crack mouth opening displacements. The nominal notch crack length to width ratio, $a_n/W$, was 0.5. Hoop weld SE(B) specimens were extracted on the weld centreline to ensure that crack growth took place only in weld material. Axial weld SE(B) specimens were aligned such that the notch lay on the weld centreline. For all fracture mechanics specimens, oversized blanks were extracted, then etched to confirm the location of the weld before each specimen was machined to its final size. Specimen drawings are provided in Appendix C.

Extraction and machining of the R(T) and SE(B) specimens took place at CNC Precision Ltd, Redditch. The test matrix detailing the type and number of specimens extracted from the pipes is shown in Appendix A.

### 3.4.2 Tensile test method

Tensile tests were carried out to BS EN ISO 6892-1:2009 (British Standards Institution 2009) using an MTS Alliance RT/100 servo-electric test machine and an MTS 634-31F-24 extensometer, at ambient temperature. The extensometer was aligned to the radial direction of the pipe, which was marked on each specimen. The exact cross-sectional area of each specimen was determined by measuring the diameter of each specimen in three places using a Vernier calliper and taking an average. The data were entered into the MTS TestWorks program so that
engineering stress values could be calculated as the test was running. The extensometer was calibrated before each test using a drum micrometer.

The crosshead displacement rate applied was 0.0011 s\(^{-1}\), equivalent to 2 mm min\(^{-1}\). Measurements were made of the 0.2\% proof strength (\(\sigma_{0.2}\) or \(R_{0.2}\)), the tensile strength (\(\sigma_{TS}\) or \(R_m\)), the elongation to fracture (\(A\)) and the reduction of area after fracture (\(Z\)).

### 3.4.3 Tensile test analysis

The engineering stress, \(\sigma_{eng}\), was calculated by dividing the applied force, \(F\), by the original cross-sectional area of the parallel length of the specimen, \(S_0\), Equation 3-23. The engineering strain, \(\varepsilon_{eng}\), was calculated using Equation 3-24, where \(L\) is the gauge length of the specimen measured by the extensometer and \(L_0\) is the original gauge length.

\[
\sigma_{eng} = \frac{F}{S_0} \quad 3-23
\]

\[
\varepsilon_{eng} = \frac{L - L_0}{L_0} \quad 3-24
\]

The test data were ordered into monotonically increasing values of strain, in increments of 0.0002. For each increment of engineering strain, the engineering stress was found through interpolation of the engineering stress values. The interpolated data were used to calculate values of true stress, \(\sigma_{true}\), using Equation 3-25. True strain, \(\varepsilon_{true}\), was calculated using Equation 3-26, and the elastic and plastic components of strain, \(\varepsilon_{el}\) and \(\varepsilon_{pl}\), using Equations 3-27 and 3-28 respectively. Parameters such as the 0.2\% proof strength (the value of stress at 0.002 plastic strain) were calculated for each test specimen.

\[
\sigma_{true} = \sigma_{eng}(1 + \varepsilon_{eng}) \quad 3-25
\]

\[
\varepsilon_{true} = \ln(1 + \varepsilon_{eng}) \quad 3-26
\]
To produce a single stress-strain curve for the parent material, the engineering stress at each value of engineering strain was averaged across all of the parent tensile test specimens, to produce a single curve. The procedure was repeated for the weld tensile specimens.

### 3.4.4 Fracture toughness test method

#### 3.4.4.1 Fatigue pre-cracking

The fracture toughness test specimens were fatigue pre-cracked in accordance with the ESIS P2-92 procedure (European Structural Integrity Society 1992). The face of each specimen, close to the notch, was lightly polished by hand to make the fatigue crack easier to observe.

Pre-cracking was conducted at ambient temperature using an Amsler Vibrophore resonance testing machine, which applied a compressive sinusoidal load to each specimen. The pre-crack was grown to a crack length-to-width ratio of $a_0/W = 0.5$. Where crack growth was uneven, specimens were rotated through 180 degrees to straighten the crack. On some specimens it was necessary to place a piece of shim under the central pin, offset towards the side of least crack growth, to encourage faster crack growth, straightening the pre-crack.

The final maximum stress intensity factor applied during the pre-cracking stages was $K_{fmax} = 22 \text{ MPa}\sqrt{\text{m}}$, with an $R$-ratio of 0.1. Side-grooves were machined into each specimen once pre-cracking had been completed.

#### 3.4.4.2 Multi-specimen fracture toughness tests

Testing was carried out at ambient temperature using the multi-specimen technique, where each test was terminated at a predetermined displacement.
according to the amount of crack growth assessed in each specimen. An Instron 1362 servo-electric test machine, using the Instron Fracture Toughness test software package was used. The specimen temperature was measured using a K-type thermocouple attached to the specimen surface close to the crack tip. Crack Mouth Opening Displacement (CMOD) was measured using a clip gauge mounted on the integral knife-edges located on the specimen. Prior to testing the clip-gauge was calibrated using a drum micrometer. The exact width of each specimen, \( W \), and the gross thickness, \( B_g \), were measured using a Vernier calliper before each test was performed. A test rate of 0.24 mm min\(^{-1}\) was applied to the hoop specimens and 0.4 mm min\(^{-1}\) for the axial specimens. Stress intensity factor rates in the elastic region were 0.51 MPa\(\sqrt{m} \) sec\(^{-1}\) and 0.84 MPa\(\sqrt{m} \) sec\(^{-1}\) respectively.

![Central pin ThermoCouple Roller Specimen Clip gauge](image)

Figure 3-24 - Specimen undergoing fracture toughness testing.

Once sufficient crack growth had been achieved in each specimen, the test was terminated. On removal from the test machine, each specimen was placed on a hot plate at approximately 280°C for 1 hour to heat-tint the fracture surface to reveal the different stages of crack growth more clearly. The specimens were then fatigue loaded to delineate the end of ductile crack growth that had occurred during testing, and were then broken open. Crack lengths were calculated using the 9-point method described in the ESIS procedure (1992), using a measuring microscope.
3.4.5 Fracture toughness test analysis

Fracture toughness calculations were carried out using a verified spreadsheet complying with the analysis methods given in ESIS P2-92 (1992). Fracture toughness was expressed in terms of $J$, corrected for crack extension, where

$$J = J_0 \left\{ 1 - \frac{(0.75\eta - 1)\Delta a}{(W - a_0)} \right\}$$  \hspace{1cm} 3-29

and

$$J_0 = \frac{\eta U}{B_n(W - a_0)}$$  \hspace{1cm} 3-30

where

$$\eta = 2 + 0.522(1 - a_0/W)$$  \hspace{1cm} 3-31

and $W =$ specimen width, $B_n =$ net specimen thickness, $\Delta a =$ crack extension, $a_0 =$ initial crack length and $U =$ total area under the force-load line displacement curve.

Three $J$-resistance curves were generated using the procedure outlined in ESIS P2-92 (1992). One curve was generated for the parent material, one for the hoop weld specimens and one for the axial weld specimens. The initiation parameters $J_{0.2}$ (the value of $J$ at 0.2 mm crack extension) and $J_{0.2BL}$ (the value of $J$ at the intersection of the curve fit with the 0.2 mm offset blunting line) were calculated. This was carried out by fitting offset power curves of the form shown in Equation 3-32, to the valid test data between the 0.1 mm offset blunting line, the upper $J$ limit, $J_{max}$, and the $\Delta a_{max}$ offset line.

$$J = m + l(\Delta a)^x$$  \hspace{1cm} 3-32

In Equation 3-32, $m$, $l$, and $x$ are curve fit coefficients with $m, l, x \geq 0$ and $x \leq 1$.

When $m = 0$ the curve becomes a standard power curve; when $x = 0$ the curve becomes linear. The blunting line was calculated from:

$$J = 3.75R_m\Delta a$$  \hspace{1cm} 3-33
Where $R_m$ is the mean tensile strength for the material at the test temperature. 

ESIS P2-92 (1992) specifies a limit, $J_{max}$, to ensure that a suitably conservative measure of fracture toughness had been obtained for the material, from the smaller of:

$$J_{max} = (W - a_0) \frac{(R_m - R_{p0.2})}{20}$$ \hspace{1cm} 3-34

or,

$$J_{max} = B \frac{(R_m - R_{p0.2})}{20}$$ \hspace{1cm} 3-35

For both the parent and weld material, Equation 3-35 set the $J_{max}$ value.

The results of the chemical composition tests, metallography, residual stress measurements, tensile tests and fracture mechanics tests are described in the following section.
4 Experimental results

4.1 Chemical composition testing

Table 4-1 and Table 4-2 detail the chemical composition of the weld and parent material respectively. Both materials generally meet the compositional requirements for austenitic stainless steel pipe, specified by the ASTM A312/A312M specification (American Society for Testing and Materials 2011a).

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cu</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>Max wt %</td>
<td>0.030</td>
<td>2.000</td>
<td>0.750</td>
<td>0.030</td>
<td>0.045</td>
<td>0.045</td>
<td>12.000</td>
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<td>Min wt %</td>
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<td></td>
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<td></td>
<td></td>
<td>10.000</td>
</tr>
<tr>
<td>Test result</td>
<td>0.012</td>
<td>1.850</td>
<td>0.310</td>
<td>0.017</td>
<td>0.011</td>
<td>0.040</td>
<td>10.000</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Element</th>
<th>Cr</th>
<th>Mo</th>
<th>Nb</th>
<th>V</th>
<th>Al</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>Max wt %</td>
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<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Min wt %</td>
<td>20.000</td>
<td></td>
<td></td>
<td></td>
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</tr>
<tr>
<td>Test result</td>
<td>20.250</td>
<td>0.020</td>
<td>0.010</td>
<td>0.050</td>
<td>0.007</td>
<td>&lt;0.010</td>
</tr>
</tbody>
</table>

Table 4-1 - Chemical composition of 308L stainless steel weld material.

<table>
<thead>
<tr>
<th>Element</th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>S</th>
<th>P</th>
<th>Cu</th>
<th>Ni</th>
</tr>
</thead>
<tbody>
<tr>
<td>Max wt %</td>
<td>0.030</td>
<td>2.000</td>
<td>1.000</td>
<td>0.030</td>
<td>0.045</td>
<td>-</td>
<td>11.000</td>
</tr>
<tr>
<td>Min wt %</td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td>8.000</td>
</tr>
<tr>
<td>Test result</td>
<td>0.026</td>
<td>1.690</td>
<td>0.270</td>
<td>0.007</td>
<td>0.021</td>
<td>0.250</td>
<td>10.030</td>
</tr>
</tbody>
</table>

<table>
<thead>
<tr>
<th>Element</th>
<th>Cr</th>
<th>Mo</th>
<th>Nb</th>
<th>V</th>
<th>Al</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>Max wt %</td>
<td>20.000</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Min wt %</td>
<td>18.000</td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Test result</td>
<td>18.000</td>
<td>0.260</td>
<td>0.010</td>
<td>0.060</td>
<td>0.011</td>
<td>&lt;0.010</td>
</tr>
</tbody>
</table>

Table 4-2 - Chemical composition of 304L stainless steel parent material.

The low carbon content of both materials reduces carbide formation, in particular chromium carbide, which can lead to sensitisation of the material microstructure. The chromium content is controlled to balance corrosion resistance with strength.
The presence of manganese increases strength, toughness and hot-working properties. The presence of sulphur, an impurity, and silicon, a deoxidising agent, result from the steelmaking process. The low concentration of these elements reflects the cleanliness of the steelmaking process used. In particular, a low sulphur concentration reduces the number of inclusions in the steel, which are potential defect initiation sites. Molybdenum is present in both the weld and parent metals, which, although not strictly to specification, is often added to stainless steels to reduce the effects of pitting corrosion, particularly in chloride environments (Higgins 1983).

4.2 Microstructural properties

4.2.1 Parent microstructure

Figure 4-1 and Figure 4-2 show the microstructure of the parent material, in the radial-axial and radial-hoop planes respectively. The two micrographs show that the microstructure is both isotropic and homogeneous. The average grain size was measured as $28 \pm 6 \, \mu m$, using the line-intercept method, indicating that the material is of a fine-grained microstructure. Observing each sample at 100x magnification, and comparing the image to grain size charts, the ASTM grain size number for the specimens ranged from 7 to 7.5, satisfying the requirements of the ASTM specification for seamless, welded and heavily cold worked austenitic stainless steel pipe, ASTM A312/A312M (American Society for Testing and Materials 2011a). This equated to grain sizes of between $27 \, \mu m$ and $32 \, \mu m$, supporting the result obtained using the line intercept method.
4.2.2 Weld microstructure

Figure 4-3 and Figure 4-4 show the microstructure of the weld material in the radial-axial and radial-hoop planes respectively. The dendritic microstructure is typical of an austenitic weld, with no grains observable using optical microscopy. Directionality is present in the weld microstructure, due to the multi-pass welding technique used. The appearance of the weld microstructure showed little variation across the width of the weldment, due to the narrow-gap profile, allowing heat to dissipate easily into the parent material. Therefore, the temperature gradient across the weld, and thus the grain growth rate, is similar across the weldment.
4.2.3 Weld-parent transition microstructure

Figure 4-5 shows the boundary between the weld and parent material microstructures. The transition region between the microstructures is too small to observe using optical microscopy, even at high magnifications, as shown in Figure 4-6. The multi-pass welding technique is evident in Figure 4-7. The small transition region is a result of the NG welding process: the relatively small quantity of filler metal requires a lower heat input, reducing the size of the HAZ (Engelhard et al. 2000).
Figure 4-5 - Micrograph of the weld and parent boundary in the radial-axial plane.

Figure 4-6 - Micrograph of the weld and parent boundary in the radial-axial plane.

Figure 4-7 - Micrograph of the weld and parent boundary in the radial-axial plane.
4.3 Tensile test results

Data from the tensile tests are shown in Table 4-3, summarising the 0.2% proof strength, tensile strength, elongation and reduction in area for each tensile specimen. The flow strength and yield to tensile strength ratio were also calculated. Table 4-4 summarises the mean value of each parameter for both materials tested. The averaged stress-strain curves of the parent and weld material are shown in Figure 4-8.

For the parent material, the mean values of 0.2% proof strength and tensile strength were 284 MPa and 592 MPa respectively. For the weld material, the mean values of 0.2% proof strength and tensile strength were 459 MPa and 605 MPa respectively. This indicates that the weld is just over 60% overmatched. The Ramberg-Osgood work hardening index, \( n \), was calculated for the parent and weld material using a method detailed in Section 5.2.4. The work-hardening behaviour of both materials is similar; at the 0.2% proof strength of the weld material, the parent material shows similar levels of work-hardening as the weld material. In Figure 4-9 the parent stress-strain curve above 459 MPa (the yield strength of the weld material) has been superimposed on the weld stress-strain curve to illustrate the similar work-hardening behaviour of the two materials. The work-hardening index for the parent material, \( n = 2.2 \), and for the weld material, \( n = 2.8 \), were similar to values obtained by Cofie et al. (1989), which ranged from \( n = 2.1 \) to \( n = 3.1 \) for stress-strain data obtained from tensile testing 304 stainless steel.

The scatter in the parent and weld tensile data was similar: the measured proof strength of the parent and weld material had a standard deviation of 16 MPa and 14 MPa respectively. Tensile specimens RT/W/6 and RT/W/7, extracted from the large pipe weld section, showed very similar stress-strain behaviour to the weld specimens extracted from the short pipe section.
<table>
<thead>
<tr>
<th>Specimen code</th>
<th>Material</th>
<th>Source pipe</th>
<th>Test temperature (°C)</th>
<th>0.2% proof strength (MPa)</th>
<th>UTS (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of area (%)</th>
<th>Proof/UTS ratio</th>
</tr>
</thead>
<tbody>
<tr>
<td>RT/P/1</td>
<td>Parent</td>
<td>Short</td>
<td>21</td>
<td>284</td>
<td>590</td>
<td>77</td>
<td>83</td>
<td>0.48</td>
</tr>
<tr>
<td>RT/P/3</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>268</td>
<td>591</td>
<td>80</td>
<td>86</td>
<td>0.45</td>
</tr>
<tr>
<td>RT/P/5</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>300</td>
<td>594</td>
<td>79</td>
<td>84</td>
<td>0.51</td>
</tr>
<tr>
<td>RT/W/1</td>
<td>Weld</td>
<td>&quot;</td>
<td>&quot;</td>
<td>475</td>
<td>618</td>
<td>52</td>
<td>65</td>
<td>0.77</td>
</tr>
<tr>
<td>RT/W/2</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>446</td>
<td>598</td>
<td>58</td>
<td>53</td>
<td>0.75</td>
</tr>
<tr>
<td>RT/W/4</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>444</td>
<td>600</td>
<td>53</td>
<td>68</td>
<td>0.74</td>
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<tr>
<td>RT/W/6</td>
<td>&quot;</td>
<td>Long</td>
<td>&quot;</td>
<td>466</td>
<td>602</td>
<td>50</td>
<td>65</td>
<td>0.77</td>
</tr>
<tr>
<td>RT/W/7</td>
<td>&quot;</td>
<td>&quot;</td>
<td>&quot;</td>
<td>465</td>
<td>606</td>
<td>57</td>
<td>64</td>
<td>0.77</td>
</tr>
</tbody>
</table>

Table 4-3 - Tensile test results for each specimen.

<table>
<thead>
<tr>
<th>Material</th>
<th>0.2% proof strength (MPa)</th>
<th>Tensile strength (MPa)</th>
<th>Flow strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of area (%)</th>
<th>Proof/TS ratio</th>
<th>Work hardening index, n</th>
</tr>
</thead>
<tbody>
<tr>
<td>Parent</td>
<td>284</td>
<td>592</td>
<td>438</td>
<td>79</td>
<td>84</td>
<td>0.48</td>
<td>2.2</td>
</tr>
<tr>
<td>Weld</td>
<td>459</td>
<td>605</td>
<td>532</td>
<td>54</td>
<td>63</td>
<td>0.76</td>
<td>2.8</td>
</tr>
</tbody>
</table>

Table 4-4 - Summary of mean parent and weld material tensile properties.

Figure 4-8 - Stress-strain curves for the parent and weld material.
Figure 4-9 - Stress-strain curves for parent and weld material. The parent material curve beyond 459 MPa is offset to overlay the weld tensile curve, to show the similar work-hardening characteristics of the two materials.

4.4 Weld residual stress measurements

4.4.1 Deep Hole Drilling

Figure 4-10 shows the through-wall residual stress profile on the weld centreline of the pipe weld, as measured using DHD. Both the hoop and axial components of stress are tensile on the outside surface of the pipe. On the weld centreline, it is clear that the residual stresses are not self-balancing; both hoop and axial stresses are predominantly tensile.

On the outer surface, the hoop residual stress is 275 MPa in tension, rising to a peak tensile stress of 375 MPa, 12 mm from the Outer Diameter (OD). The hoop stress diminishes to zero, 23 mm from the OD, beyond which the hoop stresses become compressive. The peak compressive hoop stress is 180 MPa, 28.5 mm from the OD, diminishing to zero at the inner surface.
The axial residual stress is 100 MPa in tension on the outer surface, reaching a peak of 205 MPa, 7.5 mm from the OD. The axial residual stress falls to zero, 15.5 mm from the OD. The peak tensile axial residual stress is 160 MPa, 27.5 mm from the OD, diminishing to zero 31 mm from the OD. The axial residual stresses then become tensile again: on the inner surface of the pipe, the axial residual stress is 65 MPa in tension.

Figure 4-10 - Through-wall residual stress profile, measured using DHD on the weld centreline. For clarity, error bars are not shown. Error in results: ±25 MPa.

The DHD results show that the peak tensile hoop residual stress is almost twice the peak tensile axial residual stress. The difference between peak compressive hoop and peak compressive axial residual stresses is not so large, differing by approximately 10% in magnitude.

Figure 4-11 shows the DHD data normalised by the 0.2% proof strength of the weld material. On the weld centreline, it is clear that the weld residual stress does not exceed the proof strength of the weld material. At most, the peak residual hoop stress value in tension reaches 80% of the 0.2% proof strength. In tension, the peak axial residual stress does not exceed 40% of the proof strength.

4.4.2 Neutron Diffraction

Figure 4-12 shows the residual stress field across the weldment, 10 mm below the OD, measured using Neutron Diffraction (ND). All of the components of residual stress at this depth are tensile, consistent with the DHD measurements shown in Figure 4-10. The magnitude of the hoop and axial stresses vary significantly across the weldment, showing a double peak in the area surrounding the weldment.

The hoop residual stress has a global maximum of 500 MPa, 6 mm from the weld centreline on the Left Hand Side (LHS). Two local maxima, of 470 MPa and 475 MPa, lie 3 mm and 9 mm respectively from the Right Hand Side (RHS) of the weld centreline. The difference in the LHS and RHS residual hoop stress peaks is less than 6%. At 22 mm from the centreline, the hoop residual stresses equal the axial residual stress, at a magnitude equal to the proof strength of the parent material, 284 MPa. The hoop residual stress falls to approximately half of the maximum value, 245 MPa, at 24 mm from the weld centreline. The hoop stress reduces to 140 MPa at 50 mm from the centreline, almost a quarter of the maximum hoop stress.
The axial residual stresses follow a similar trend to the hoop stresses. A global maximum of 440 MPa is situated 6 mm from the centreline on the LHS. A local maximum of 430 MPa is positioned 9 mm on the RHS of the centreline. The two maxima have magnitudes within 2% of each other. The axial residual stresses do not diminish from the weld centreline as quickly as the hoop residual stresses. The axial stresses are higher than the hoop stresses from 22 mm to 49 mm from the centreline. 24 mm from the centreline, the axial stress is approximately two-thirds of its peak value, 285 MPa. At the same distance, the hoop stresses are half their peak value. At 49 mm from the centreline, the axial residual stress is 240 MPa.

The radial residual stresses show less variation across the weldment when compared to the axial and hoop stresses. Two maxima can be found in the same places as for the axial and hoop stresses: 6 mm from the centreline on the LHS and 9 mm from the centreline on the RHS, 240 MPa and 235 MPa respectively. A minimum is situated 3 mm from the LHS of the centreline, similar to the hoop and axial stresses. The radial stresses remain lower than the axial stresses across the weldment. However, the radial stress is approximately equal to the hoop stress,
230 MPa, 30 mm from the centreline on the LHS. Beyond 30 mm, the radial stress exceeds the hoop stress. At 49 mm from the centreline, the radial residual stress is 205 MPa, only 16% less than the peak radial stress.

The points at 0 mm and 3 mm lie in the weld material. At 6 mm, the gauge volume lies on the boundary of the weld and parent material. Beyond 6 mm, the gauge volume lies in parent material. Figure 4-12 shows that the peak values of all three components of residual stress lie outside the weldment. The axial and hoop residual stresses exceed the yield strength of the parent material up to 2 mm either side of the weld centreline. Only the hoop stresses exceed the yield strength of the weld material, 459 MPa, up to 9 mm from the centreline. At high residual stresses, there is a discrepancy between the DHD and the ND results: for example, on the weld centreline, at 10 mm depth, the residual hoop stress measured by DHD is 345 MPa, whereas the ND measurement is 435 MPa, a difference of 23%. This discrepancy is discussed in more detail in Section 6.

The ND residual stress data show a degree of symmetry. The local maxima of all three components of stress occur at the same distance from the weld centreline. Local minima also occur in similar locations, similar to the results of Liu et al. (2011) and Yang et al. (2011). To illustrate this more clearly, the data used to plot Figure 4-13 were modified. The average of the two local maxima, at -6 mm and 9 mm, was calculated for each component of residual stress. Similarly, the average of the local minima, at -3 mm, 0 mm and 3 mm, was calculated for each residual stress component. Plotting the average of the local minima and maxima revealed a more symmetrical residual stress profile, shown in Figure 4-13; a comparison of the residual stress profile with published data is made in Section 6.
Figure 4-13 shows that all three residual stress components have a centre of symmetry lying at approximately 1.5 mm to the right of the weld centreline. The stress profile is not perfectly symmetric; this could be due to the particular configuration of the weld passes where the linescan was performed. Performing similar linescans in different areas of the pipe, and averaging the results, may have revealed a greater degree of symmetry, centred on the weld centreline.

Figure 4-14 shows the through-wall residual stresses on the weld centreline, measured by ND. Due to the low uncertainty in the ND measurements, error bars are too small to display. However, the measurement made at 15 mm depth may not be representative of the true residual stress at that point. The depth at which the measurement was made caused a high level of signal attenuation, leading to a poor Signal-to-Noise Ratio. The resultant $\sin^2\psi$ plot, shown in Figure 4-15, did not show an ellipsoidal shape, leading to a poor fit from which hoop and radial strains were calculated. Therefore the measurement made at 15 mm depth may be considered unrepresentative of the true residual stress at that depth. However, the axial stress measurements proved more straightforward; the short path length permitted direct measurements of axial strain at all depths, without resorting to curve fitting.
general, the through-thickness ND measurements show a similar form to the DHD results, with peak tensile hoop and axial residual stresses located at approximately 10 mm from the OD. The results obtained using the two methods are compared in more detail in Section 6.

Figure 4-14 - Through-wall residual stress profile, comparing measurements using ND.

Figure 4-15 – \( d \) vs. \( \sin^2 \psi \) plot measured at 15 mm depth.
4.5 Retained residual stress measurements

4.5.1 Hoop-orientated SE(B) blank specimen

Figure 4-16 shows the hoop and axial components of residual stress measured in the blank hoop specimen, HP/HD1. The solid lines indicate the level of residual stress partially relaxed upon extraction of the blank SE(B) specimen. The dashed lines indicate the fully relaxed residual stresses upon wire-EDM trepanning. The difference in stress between the dashed and solid lines is equal to the residual stress retained in the blank SE(B) specimen. Figure 4-16 shows that there was little difference in the residual stress measured over the two stages, indicating that there was little residual stress retained in the hoop-orientated specimen. This is illustrated more clearly in Figure 4-17.

Figure 4-16 - Residual stresses measured during the extraction of the hoop-orientated blank SE(B) specimen, HP/HD1. Error in results: ±25 MPa.
Figure 4-16 shows that the hoop component of the original weld residual stress field had a peak tensile stress of 485 MPa, 9 mm from the OD. The hoop stress diminished to zero, 20 mm from the OD, beyond which the hoop stresses became compressive. The peak compressive hoop stress was 260 MPa, 27 mm from the OD, remaining compressive at the inner surface.

The axial component of the original weld residual stress field reached a peak of 340 MPa, 9 mm from the OD. The axial residual stress fell to zero, 15 mm from the OD. The peak tensile axial residual stress was 290 MPa, 28 mm from the OD, diminishing to zero 21 mm from the OD. The axial residual stresses became tensile at the inner surface of the pipe.

Figure 4-17 shows the retained hoop and axial residual stresses in the hoop-orientated SE(B) specimen. It is clear that most of the residual stress had relaxed upon extraction of the specimen from the pipe weld. The peak tensile hoop and axial stresses were 80 MPa and 75 MPa respectively. The peak compressive hoop and axial stresses were 110 MPa and 95 MPa respectively.

Figure 4-17 - Residual stress retained in hoop-orientated SE(B) specimen HP/HD1. Error in results: ±25 MPa.
**4.5.2 Axially-orientated SE(B) blank specimen**

Figure 4-18 shows that a significant level of residual stress was retained in the axially-orientated SE(B) specimen, AX/HD1. This is illustrated more clearly in Figure 4-19, which shows that the peak tensile hoop and axial stress retained in the specimen was 110 MPa and 250 MPa respectively. The peak compressive hoop and axial stress retained in the specimen was 40 MPa and 140 MPa respectively. Of particular interest is the peak tensile axial stress of 250 MPa, equal to 54% of the 0.2% proof strength of the weld material. This shows that a significant level of retained residual stress may make a contribution to the crack opening forces during a fracture toughness test; this is discussed in more detail in Section 6.

![Residual stresses measured during the extraction of the axially-orientated blank SE(B) specimen, AX/HD1. Error in results: ±25 MPa.](image)

Figure 4-18 - Residual stresses measured during the extraction of the axially-orientated blank SE(B) specimen, AX/HD1. Error in results: ±25 MPa.
4.5.3 Comparison with original DHD measurements

Table 4-5 compares the measurements of the original pipe weld residual stress field, obtained from the original DHD measurement and the DHD performed on the blank SE(B) specimens. The peak residual stress values, and the depth where they lie, is consistent between the DHD measurements performed on the blank SE(B) specimens. However, there is a discrepancy between the SE(B) measurements and the original DHD measurement. An explanation for this discrepancy, arising from the different DHD methods used in each case, is discussed in detail in Section 6.
Table 4-5 - Comparison of residual stress measurements obtained from original DHD data and SE(B) DHD data.

### 4.6 Fracture toughness test results

Five fracture toughness tests were carried out using hoop-orientated weld specimens, and six tests were carried out for the axial-orientated weld specimens. The measurement data are listed in Appendix A. All tests, except specimens W2 and AX7, met the requirement for all nine crack front measurements to be within 10% of B, as set out in ESIS P2-92. Specimens W2 and AX7 were excluded from the J-resistance curve calculations due to uneven pre-cracks. The fracture toughness data calculated from the measured end points for all specimens are summarised in Appendix A.

For the parent material fracture toughness tests, an offset power law curve fit, Equation 3-32, yielded a negative value of m, so a power law curve was fitted to the data instead (where m = 0). \( J_{0.2BL} \) could not be reliably established, as this would have involved extrapolating the J-resistance curve beyond the range of experimental data.

A power law curve fit was applied to the hoop-orientated weld specimen data, because it was not possible to reliably fit an offset power law curve to the test data. An offset power law curve was successfully fitted to the axially-orientated weld specimens. A summary of the fit parameters, and respective correlation coefficients,
is included in Appendix A. Table 4-6 summarises the fracture toughness measurements of the two materials and their respective orientations, using the two curve-fitting approaches.

<table>
<thead>
<tr>
<th>Specimen material/orientation</th>
<th>Offset power-law curve fit</th>
<th>Power-law curve fit</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$J_{0.2}$ (kJ/m$^2$)</td>
<td>$J_{0.2BL}$ (kJ/m$^2$)</td>
</tr>
<tr>
<td>Parent, hoop</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Weld, hoop</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Weld, axial</td>
<td>387</td>
<td>531</td>
</tr>
<tr>
<td>Weld, combined</td>
<td>395</td>
<td>521</td>
</tr>
</tbody>
</table>

Table 4-6 - Summary of fracture toughness parameters using two curve fits.

Table 4-7 summarises the fracture toughness measurements in terms of $K_J$.

<table>
<thead>
<tr>
<th>Specimen material/orientation</th>
<th>Offset power-law curve fit</th>
<th>Power-law curve fit</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$K_{J0.2}$ (MPa√m)</td>
<td>$K_{J0.2BL}$ (MPa√m)</td>
</tr>
<tr>
<td>Parent, hoop</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Weld, hoop</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Weld, axial</td>
<td>290</td>
<td>340</td>
</tr>
<tr>
<td>Weld, combined</td>
<td>293</td>
<td>337</td>
</tr>
</tbody>
</table>

Table 4-7 - Summary of fracture toughness parameters, converted to $K_J$ values.

Figure 4-20 shows the $J$-resistance curve for the parent material, Figure 4-21 shows the $J$-resistance curve for the hoop-orientated weld specimens and Figure 4-22 shows the $J$-resistance curve for the axially-orientated weld specimens.
Figure 4-20 - $J$-resistance curve for the parent material.

Figure 4-21 - $J$-resistance curve for the weld material, using hoop-orientated specimens.
All of the curve fits showed good correlation coefficient values between $R = 0.975$ and $R = 0.999$. However, the data points from all of the tested specimens lay well above the $J_{max}$ limit set by ESIS P2-92. This suggests that a sufficiently conservative measure of fracture toughness may not have been obtained, due to the relatively small thickness (10 mm) of the SE(B) specimens, even with side-grooves present. However, it was not possible to extract larger specimens from the narrow-gap weld without encroaching on neighbouring parent material.

The data for the axial and hoop orientated weld specimens were similar, so the two datasets were combined into a single $J$ vs. $\Delta a$ plot, shown in Figure 4-23. This approach had several advantages: combining the data provided more data points, from which a more reliable curve fit could be made, and combining the two orientations sampled a wider range of weld properties. Both an offset power law curve and a power law curve were fitted to the data. When comparing the $J$-resistance curves calculated using the power law approach, it can be seen, from Table 4-6, that $J_{0.2}$ from the combined data lies halfway between $J_{0.2}$ calculated for the separate axial and hoop curves. This may suggest that the weld fracture toughness is approximately the same for both orientations. However, this cannot be
confirmed due to the relatively small amount of data available. Further fracture toughness tests would be needed for a more conclusive result. It must also be noted that the $J$-resistance curve for the combined data lies well above the $J_{\text{max}}$ validity limit for the specimen sizes tested. Further discussion of the measured fracture toughness, and the validity of the results, is discussed in Section 6.

Figure 4-23 - $J$-resistance curve for the weld material, combining the data from the hoop and axially-orientated specimens.
5 Numerical modelling

5.1 Introduction

A numerical model was developed to simulate the structural behaviour of the narrow-gap pipe weld with a postulated defect present. The Finite Element Analysis (FEA) software package, Abaqus v6.8 (Dassault Systemes Simulia Corporation 2008), was used. Different approaches to establish the Crack Driving Force (CDF) were compared. One approach used full elastic-plastic cracked body analysis, simulating the behaviour of the pipe weld under service loading. A second approach calculated the CDF for a component under combined loading using the current R6 methodology, through the use of the $\rho$-interaction parameter. Numerical modelling provided the opportunity to utilise the $g$-parameter (James et al. 2009), aimed at reducing the conservatism of the current $\rho$-parameter based approach. CDF predictions based on the two approaches were compared.

This chapter describes the development of the FE model and its application to a structural integrity assessment. The numerical model was written using a series of text-based input files, giving precise control of the model geometry, mesh refinement and allocation of node and element numbers. A fully-circumferential, 10 mm deep crack was modelled. Two loading modes were applied: a global, tensile axial load and a pressure load. The more straightforward nature of modelling a tensile axial load was used during development of the input files. The axial load models were then modified to simulate pressure loading, which was more representative of in-service loading. A comparison between the CDF predictions under axial and pressure loading was made.
5.2 Modelling approach

5.2.1 Uncracked elastic model

The pipe weld was modelled using two-dimensional (2-D), 8-node, second-order axially symmetric elements with reduced integration (CAX8R). Figure 5-1 shows the 2-D mesh within the context of the modelled component.

![2-D FE mesh](image1)

Figure 5-1 - 2-D FE mesh (shown in green), within the context of the pipe weld.

66 elements, each 0.5 mm in length, ran through the pipe wall. 40 elements ran along the axis of the model, biased to increase the density of the elements close to the weld centreline. Figure 5-2 shows the mesh in more detail.

![Axisymmetric FE mesh](image2)

Figure 5-2 - Axisymmetric FE mesh used to model pipe weld.
A boundary condition was applied to the nodes running along the weld centreline, constraining the nodes to prevent motion in the axial direction. An increasing tensile axial load in the pipe wall was simulated by applying a concentrated load to a master node, which was tied to nodes on the bottom edge of the model to enforce equal displacements. The mesh is shown in Figure 5-3. The elastic material properties of the parent material were applied to the model: a Young’s Modulus, \( E \), of 198 GPa and a Poisson’s ratio, \( v \), of 0.3.

![FE mesh showing the application of a boundary condition on the weld centreline, preventing motion in the axial direction, and the application of equal displacements to the base of the model.](image)

To model the weld residual stresses, the uncracked elastic model was used to develop the application of a temperature distribution across the weld centreline. By adjusting the temperature levels across the model, different levels of strain were introduced to the material through thermal expansion and contraction, simulating residual stresses across the weldment. A user-defined subroutine was developed using Fortran code to create the temperature distribution. Along the radial axis, the distribution took a sinusoidal form, shown in Equation 5-1.
\[ T_x = A \left( 1 + \cos \left( \frac{\pi x}{t} \right) - (\phi) \right) \]

\( T_x \) is the temperature along the radial axis of the model (along the weld centreline), \( x \) is the distance across the weld centreline and \( t \) is the thickness of the pipe wall. The terms \( A \) and \( \phi \) were the amplitude and phase terms respectively.

A second function was developed to model the decay of weld residual stress away from the weld centreline in the axial direction. An exponential function was used, shown in Equation 5-2, where \( T_y \) is the temperature along the pipe axis, \( y \) is the distance away from the weld centreline and \( \mu \) is the variable controlling the rate of decay.

\[ T_y = e^{-\frac{y^2}{\mu}} \]

The resultant residual stresses were analysed using Abaqus Viewer. The modelled residual stresses were compared to the residual stresses measured by DHD. The relatively simple nature of the FE model could not fully replicate the stress field set up by the complex welding process that was used in manufacturing the real component. Matching the modelled residual stresses was achieved through trial and improvement, by adjusting the amplitude term, \( A \), the phase term, \( \phi \) and the decay term, \( \mu \). For a circumferential crack, the dominant crack opening forces caused by the residual stress field were due to the axial stresses; the temperature distribution was adjusted so that the axial stress field most closely agreed with the axial residual stresses measured using DHD. Once a satisfactory match was obtained, the optimum temperature distribution was applied to the uncracked elastic model. The temperature distribution used in the elastic-plastic FE models is described in more detail in Section 5.2.4.2; the through-wall residual stress is shown in Figure 5-10.

### 5.2.2 Cracked elastic model

For cracked body analyses, a fully-circumferential external crack, 10 mm in depth, was modelled by removing the boundary condition from the nodes on the weld centreline, up to the crack tip node, shown in Figure 5-4. A 10 mm deep crack was
modelled so that the crack tip lay in the zone of peak tensile residual stress, to produce the maximum possible secondary CDF. The crack tip elements were the same as those used in the rest of the model: square section, axially symmetric 8-node elements with reduced integration, 0.5 mm in length. Because the primary purpose of the model was to determine the elastic-plastic crack driving force, $J$, at far-field values, the mesh refinement was considered adequate. This is discussed in more detail below.

![Figure 5-4 - FE mesh showing the application of a boundary condition on the weld centreline to model a circumferential crack.](image)

A distributed load, derived from the stresses set up by the temperature distribution, was applied to the unconstrained nodes on the crack flank, shown in Figure 5-5. The crack driving force due to the applied secondary stress field, $J^S$, was calculated using Abaqus. The crack driving force, $J^S$, due to the residual stress field was converted to its elastic equivalent, $K_f^S$, using Equation 5-3.

$$K_f^S = \sqrt{\frac{E J^S}{(1 - v^2)}}$$  \hspace{1cm} 5-3
Figure 5-5 - Distributed load applied to the crack flank to model weld residual stresses.

Figure 5-6 shows the $J$ values calculated using different contours around the crack tip. Several contours were used, the first starting at the first node from the crack tip. Beyond the third contour from the crack tip, the $J$-values became path-independent. As a result, $J$ was calculated from the eighteenth contour, to ensure that $J$ remained path-independent over the range of applied loads.

Figure 5-6 - Path dependence of the $J$-integral, calculated using different contours around the crack tip.
To obtain the elastic SIF due to primary loading, $K_{s}^{p}$, the distributed load due to secondary loading was removed. An arbitrary distributed load of 100 MPa was applied to the node set running along the base of the model, shown in Figure 5-4.

5.2.3 Elastic-perfectly plastic cracked model

To obtain the plastic limit load, $P_L$, of the cracked component, a material model with elastic-perfectly plastic properties was used. The Young’s Modulus, $E$, and Poisson’s ratio, $v$, of 198 GPa and 0.3 respectively were used to model elastic properties. To model plastic behaviour, the Abaqus input file script was modified, setting the strain at the 0.2% proof strength of the parent material to zero. This was followed by a line in the script to set the strain at the proof strength to 500%, to model perfectly plastic behaviour. Isotropic hardening was applied.

A displacement of 40 mm was applied in the axial direction to the nodes on the bottom edge of the model. The Crack Mouth Opening Displacement (CMOD) was recorded using a monitor node at the crack mouth. The plastic limit load was chosen as the load at which the CMOD equalled half of the crack length (5 mm). This approach was verified by viewing the deformed model in Abaqus Viewer. When the stress contour representing the yield stress of 284 MPa was seen to extend through the full wall thickness, it was clear that the plastic zone had extended through the uncracked section of the pipe wall.

5.2.4 Elastic-plastic models

5.2.4.1 Elastic-plastic material model

The elastic-plastic material model was based upon the raw stress-strain data obtained from the tensile tests. An averaged stress-strain curve was produced from the three parent material tensile tests. The plastic strain, $\varepsilon_{pl}$, and corresponding value of true stress, $\sigma_{true}$, were ordered into monotonically increasing values of plastic strain, in increments of 0.0002, as described in Chapter 3.
Figure 5-7 shows the stress-strain data obtained from the tensile test at small strains, at the early stages of the tensile test. It can be seen that for small strains, the data lay above the elastic line, then converged with the elastic line, then diverged as plasticity developed. The scatter in the measurements made it hard to establish the boundary between the elastic region and the elastic-plastic region.

The data was modified so that the transition from elastic to elastic-plastic behaviour was smooth and clearly defined, so that the gradual increase in plastic strain with increasing stress resembled a model ductile material more closely. To clearly define an elastic region, a linear fit was made between the origin and the first data point that coincided with the elastic line. This ignored the first few data points of each tensile test where the measured strains were too small to be considered accurate. Beyond the linear fit, a 13-point average of the points surrounding each raw data point was taken to produce a smooth curve. This curve, plotted alongside a linear elastic line, is shown in Figure 5-8. It can be seen that the fitted linear region of the modified curve aligned with the elastic line very closely, beyond which increasing stress led to a gradual increase in plastic strain. This modified curve was applied to the elastic-plastic FE analyses.
To model the very high strains at the crack tip, the Ramberg-Osgood relation, Equation 2-6, was fitted to the last 30 data points of the averaged stress-strain curve, to extrapolate the tensile data up to 200% strain.

To obtain the value of $n$ for the parent material, Ramberg and Osgood show that by introducing the work-hardening coefficient, $\alpha$, shown in Equation 5-4, and substituting into Equation 2-6, the Ramberg-Osgood relationship can be expressed as shown in Equation 5-5. For the parent material, $\sigma_0$ was set equal to the 0.2% proof strength (284 MPa).

$$\alpha = K \left( \frac{\sigma_0}{E} \right)^{n-1}$$ \hspace{1cm} 5-4

$$\varepsilon = \frac{\sigma_{true}}{E} + \frac{\alpha}{E} \left( \frac{\sigma_{true}}{\sigma_0} \right)^n$$ \hspace{1cm} 5-5

The first and second terms in Equation 5-5 represent the elastic and plastic components of strain, $\varepsilon_{el}$ and $\varepsilon_{pl}$ respectively. To obtain $\alpha$ and $n$, the plastic component of strain, Equation 5-6, was considered.
Equation 5-6 was simplified by introducing a new constant, \( \chi \), Equation 5-7.

\[
\chi = \frac{a\sigma_0}{E\sigma_0^n}
\]

5-7

Which simplified to Equation 5-8.

\[
\varepsilon_{pl} = \chi \sigma_{true}^n
\]

5-8

Taking natural logarithms of Equation 5-8 yielded Equation 5-9.

\[
\ln \varepsilon_{pl} = \ln \chi + n \ln \sigma_{true}
\]

5-9

The linear relationship of Equation 5-9 was plotted as a graph of \( \ln \varepsilon_{pl} \) versus \( \ln \sigma_{true} \), the gradient of which gave \( n \), and the intercept with the y-axis the value of \( \ln \chi \). Taking the exponent of \( \ln \chi \) and rearranging Equation 5-8 yielded the value of \( a \). Values of \( n \) and \( a \) equal to 2.2 and 24 respectively, were calculated. Thus, the true stress-plastic strain data of the parent material were extrapolated to 200% strain, shown in Figure 5-9.
5.2.4.2 Uncracked elastic-plastic model

The uncracked elastic model, described in Section 5.2.2, was modified to incorporate elastic-plastic material behaviour. The temperature distribution, which had previously been applied to the elastic model, was applied to the elastic-plastic model. The modelled residual stress field was compared with the measured residual stress field obtained through DHD. The stress field was different from that created in the elastic model, due to plasticity relaxing the strains that had been set up by the temperature distribution. The temperature distribution was adjusted so that the residual stress field matched the DHD measurements as closely as possible. The final temperature distribution used is defined by Equation 5-10 and Equation 5-11 and compared with the measured axial residual stress distribution in Figure 5-10.

\[
T_x = 3.5 \left( 1 + \cos \left( \frac{\pi x}{t} \right) - \left( \frac{15\pi}{2} \right) \right)
\]

\[
T_y = e^{-\frac{y^2}{200}}
\]

Figure 5-9 - True stress-plastic strain behaviour of parent material, extrapolated to 200% strain using a Ramberg-Osgood fit.
5.2.5 Cracked elastic-plastic model

The cracked elastic-plastic model was based on the cracked elastic model, with elastic-plastic material properties applied. The modified temperature distribution was applied across the weld centreline. The circumferential crack was then introduced by removing the boundary condition, as described in Section 5.2.2. Finally, an increasing, tensile axial load was applied to the bottom node set. The applied load was scaled from $L_r = 0$ to $L_r = 1.6$, to obtain the CDF for high levels of plasticity. Results for the elastic-plastic parameter $f^{p+S}$ were obtained as a function of primary load. The crack driving force parameter for combined loading, $K_j^{p+S}$, was determined through Equation 2-36.

$$K_j^{p+S} = \frac{E f^{p+S}}{\sqrt{(1 - u^2)}}$$  \hspace{1cm} 5-12

The elastic-plastic cracked model was then modified to calculate the CDF due to primary loading only. The temperature distribution was commented out of the Abaqus input file, to remove the residual stress field. The same increasing tensile axial primary load was applied, as described above. The elastic plastic parameter
due to primary loading, $J^p$, was obtained as a function of primary load, from which the crack driving force due to primary loading only, $K^p_f$, was obtained.

### 5.3 Pressure loading

The numerical models were adapted to model the application of a pressure load to the pipe weld. The crack opening stresses in the elastic regime were purely axial, therefore the elastic CDF due to pressure loading was the same as for axial loading. As a result, the axially-loaded elastic models described above were used to calculate the elastic CDF for pressure loading.

Where plasticity was present, such as in the elastic-plastic and elastic-perfectly plastic models, it was necessary to adapt the axially loaded models to simulate pressure loading. This was achieved by applying a distributed load to the inner wall of the pipe, as well as at the base of the pipe, Figure 5-11.

![Figure 5-11 - Cracked FE model showing the application of distributed loads to the inner pipe wall and pipe end to simulate pressure loading.](image)
To calculate the correct ratio of distributed loads to apply to the model, the applied pressure, \( P \), and the axial stress, \( \sigma_{ax} \), were derived using the results from the axially-loaded models. The axial limit load of the pipe section was established using limit load analysis, described in Section 5.2.3. To obtain the axial stress at the limit load, Equation 5-13 was used.

\[
\sigma = \frac{F}{A} \quad 5-13
\]

Substituting in the force due to the applied limit load, \( F_{ax} \), and the cross-sectional area of the pipe into Equation 5-13, led to Equation 5-14, the stress equivalent to the limit load.

\[
\sigma_{ax} = \frac{F_{ax}}{\pi(r_{out}^2 - r_{in}^2)} \quad 5-14
\]

Where \( r_{out} \) and \( r_{in} \) are the outer and inner radii of the pipe, respectively.

To obtain the limit pressure, it was necessary to introduce a second expression. For a closed pipe under internal pressure, \( P \), the axial component of stress can be expressed as Equation 5-15.

\[
\sigma_{ax} = \frac{r_{in}P}{2t} \quad 5-15
\]

Where \( t \) is the wall thickness of the pipe. Substituting Equation 5-15 into Equation 5-14 yields Equation 5-16.

\[
\frac{Pr_{in}}{2t} = \frac{F_{ax}}{\pi(r_{out}^2 - r_{in}^2)} \quad 5-16
\]

Rearranging Equation 5-16 leads to Equation 5-17, which gives the corresponding pressure.

\[
P = \frac{2tf_{ax}}{\pi r_{in}(r_{out}^2 - r_{in}^2)} \quad 5-17
\]
Therefore Equation 5-14 and Equation 5-17 made it possible to estimate the ratio of the distributed loads to apply to the inner wall and end of the pipe to model pressure loading.

A Riks analysis was performed on the elastic-perfectly plastic cracked pipe model, with distributed loads applied to the model as shown in Figure 5-11. The load proportionality factor was recorded. Multiplying the proportionality factor by the applied distributed loads gave the limit pressure of the cracked pipe. For a fully circumferential crack, 10 mm deep, the limit pressure was calculated to be 65 MPa, over four times the normal operating pressure of the primary circuit of a PWR.

The elastic-plastic cracked models described in Section 5.2.5 were modified to incorporate pressure loading. The applied pressure load was scaled from $L_r = 0$ to $L_r = 1.6$, to obtain the CDF at high levels of plasticity. The elastic-plastic parameter for combined loading, $J^{P+S}$, and primary loading only, $J^P$, were obtained as a function of primary load. These parameters were converted to crack driving force parameters for combined loading, $K_f^{P+S}$, and primary loading $K_f^P$, using Equation 5-12.

5.3.1 Limit pressure validation

The limit pressure, $P_{\text{press}}^L$, calculated through FEA, was compared with the limit pressure for a fully external circumferential defect, in a thick walled cylinder, under internal pressure, defined in R6 Chapter IV.1.8.3 (2000).
Figure 5-12 - Geometry of thick-walled cylinder, containing a fully-circumferential defect, under internal pressure.

$R_i$ is the inner radius, $R_o$ is the outer radius, $R_m$ is the mean radius, $P$ the internal pressure, $a$ the crack depth and $t$ the wall thickness of the pipe.

The Tresca solution for the limit pressure, as expressed in R6, is shown in Equation 5-18.

$$\frac{p_{L,\text{press}}}{\sigma_y} = \ln \left( \frac{1 + \frac{\eta}{2} - \alpha \eta}{1 - \frac{\eta}{2}} \right) + \frac{1}{2} \left[ 1 - \left( \frac{1 - \frac{\eta}{2}}{1 + \frac{\eta}{2} - \alpha \eta} \right)^2 \right]$$

5-18

where,

$$\alpha = \frac{a}{t}$$

5-19

and,

$$\eta = \frac{t}{R_m}$$

5-20
However, if Equation 5-21 is not satisfied,

\[
\ln \left( \frac{1 + \frac{\eta}{2} - \alpha \eta}{1 - \frac{\eta}{2}} \right) > \frac{1}{2} \left[ 1 - \left( \frac{1 - \frac{\eta}{2}}{1 + \frac{\eta}{2} - \alpha \eta} \right)^2 \right] \tag{5-21}
\]
	hen the definition in Equation 5-22 is used instead.

\[
\frac{p^\text{press}_L}{\sigma_y} = \ln \left( \frac{1 + \frac{\eta}{2}}{1 - \frac{\eta}{2}} \right) \tag{5-22}
\]

In the case of the modelled pipe weld, the limit pressure was calculated using Equation 5-22. The result was multiplied by \( \frac{2}{\sqrt{3}} \) to obtain the von Mises solution, comparable with the numerical result. The FEA solution and the R6 solution agreed to within 1.5% of each other, as shown in Table 5-1.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>FEA solution</th>
<th>R6 solution</th>
<th>% difference</th>
</tr>
</thead>
<tbody>
<tr>
<td>Limit Pressure</td>
<td>65 MPa</td>
<td>66 MPa</td>
<td>1.5</td>
</tr>
</tbody>
</table>

Table 5-1 - Summary of handbook and FEA limit pressures.

5.3.2 Stress Intensity Factor validation

The elastic Stress Intensity Factors \( K_i^P \) and \( K_i^S \) calculated using FEA, were compared to the solution given in R6 Chapter IV.3.4.8 (2000), shown in Equation 5-23. The terms are related to those shown in Figure 5-13, where \( u \) is a path parallel to the radial axis, starting at the outer radius and ending at the inner radius. The other terms are the same as those defined in Figure 5-12.

\[
K_i = \frac{1}{\sqrt{2\pi a}} \int_0^a \sigma(u) \sum_{i=1}^{i=3} f_i \left( \frac{a}{l}, \frac{R_i}{l} \right) \left( 1 - \frac{u_i}{a} \right)^{i-3/2} du \tag{5-23}
\]
The term \( f_i \left( \frac{a}{t}, \frac{R_i}{t} \right) \) was expanded using look-up tables provided in the R6 procedure. The look-up table for \( \frac{a}{t} = 0.3 \) and \( \frac{R_i}{t} = 4 \) was used, as these values were closest to the dimensions of the modelled pipe. From the look-up tables, the following values of \( f \) were obtained:

\[
\begin{align*}
f_1 &= 2.000 \\
f_2 &= 2.904 \\
f_3 &= 0.535
\end{align*}
\]

Equation 5-23 therefore became Equation 5-24.

\[
K_i = \frac{1}{\sqrt{2\pi a}} \int_0^a \sigma(u) \left[ 2 \left( 1 - \frac{u}{a} \right)^{1-3/2} + 2.904 \left( 1 - \frac{u}{a} \right)^{2-3/2} + 0.535 \left( 1 - \frac{u}{a} \right)^{3-3/2} \right] du \quad 5-24
\]

The stress as a function of \( u, \sigma(u) \), was evaluated in two ways. For the SIF due to primary loading only, \( K_i^P \), an arbitrary stress distribution of 100 MPa across the length of the crack flank was chosen to evaluate Equation 5-24. For the SIF due to secondary loading, \( K_i^S \), the distributed load used in the elastic cracked model was applied.
To obtain the analytical solution for Equation 5-24 is non-trivial; numerical integration was performed instead. To obtain an accurate result, the increment $du$ was set to a small value, $du = 0.0001$. The FEA and R6 solutions are shown in Table 5-2; both solutions are in agreement to less than 1 MPa√m.

<table>
<thead>
<tr>
<th>Parameter</th>
<th>FEA solution</th>
<th>R6 solution</th>
</tr>
</thead>
<tbody>
<tr>
<td>$K^P_I$</td>
<td>25 MPa√m</td>
<td>25 MPa√m</td>
</tr>
<tr>
<td>$K^S_I$</td>
<td>46 MPa√m</td>
<td>46 MPa√m</td>
</tr>
</tbody>
</table>

Table 5-2 - Summary of handbook and FEA Stress Intensity Factors.

5.4 R6 assessment

5.4.1 Primary loading only

The CDF due to primary loading only, $K^P_I$, was determined by applying a load to the elastic-plastic cracked model, without a residual stress present. Both axial and pressure loading were modelled. $K^P_I$ was plotted against the normalised load parameter, $L_r$.

5.4.2 Combined loading using the $\rho$-parameter

$K^S_I$ was derived from the CDF due to the modelled residual stress field being applied as a distributed load to the elastic cracked model. $K^P_I$ was determined by applying an arbitrary distributed load of 100 MPa in the axial direction on the elastic cracked model.

To calculate the CDF due to combined loading using the $\rho$-parameter, the applied primary load was scaled over the range $L_r = 0$ to $L_r = 1.6$. $K^P_I$ was scaled accordingly. The ratio in Equation 5-25 was used to determine the parameters $\psi$ and $\phi$ using the look-up tables provided in R6 to obtain the value of $\rho$ at each increment of $L_r$. 
The CDF due to combined loading, $K_f^{p+s}$, was calculated using Equation 5-26.

$$\frac{K_f^s L_r}{K_f^p}$$ 

5-25

$$K_f^{p+s} = \frac{K_f^p + K_f^s}{f_3(L_r) - \rho}$$ 

5-26

Where

$$f_3(L_r) = \frac{K_i^p}{K_f^p}$$ 

5-27

### 5.4.3 Option 2 analyses

The Option 2 FAC was calculated by establishing the reference stress, $\sigma_{ref}$, using Equation 5-28.

$$\sigma_{ref} = L_r \sigma_y$$ 

5-28

The reference strain, $\varepsilon_{ref}$, was defined as the true strain corresponding to the reference stress on the true stress-true strain curve. $\varepsilon_{ref}$ was used to evaluate Equation 5-29, to produce the Option 2 FAC for the material model used.

$$f_2(L_r) = \left[ \frac{E \varepsilon_{ref}}{L_r \sigma_y + \frac{L_r^3 \sigma_y}{2E \varepsilon_{ref}}} \right]^{-1/2}$$ 

5-29

### 5.4.4 Combined loading using the $g$-parameter

To determine the CDF using the $g$-parameter (James et al. 2009), the modified reference stress, $\sigma_{ref}^{mod}$, was calculated using Equation 5-30.

$$\sigma_{ref}^{mod} = \frac{\sigma_{ref}}{1.25}$$ 

5-30
The modified reference strain, $\varepsilon_{\text{ref}}^{\text{mod}}$, was the true strain corresponding to the modified reference stress on the true stress-true strain curve. Thus, $g$ was determined using Equation 5-31.

$$
\begin{align*}
g &= \left[\frac{\sigma_{\text{ref}}^{\text{mod}}}{\varepsilon_{\text{ref}}^{\text{mod}} E} \cdot \frac{0.8 \left(\frac{\sigma_{\text{ref}}^{\text{mod}}}{\sigma_y}\right)}{\left(\frac{\sigma_{\text{ref}}^{\text{mod}}}{\varepsilon_{\text{ref}}^{\text{mod}} E}\right)}\right]^{-1/2} \\
&= \left[\frac{\sigma_{\text{ref}}^{\text{mod}}}{\varepsilon_{\text{ref}}^{\text{mod}} E} \cdot \frac{0.8 \left(\frac{\sigma_{\text{ref}}^{\text{mod}}}{\sigma_y}\right)}{\left(\frac{\sigma_{\text{ref}}^{\text{mod}}}{\varepsilon_{\text{ref}}^{\text{mod}} E}\right)}\right]^{-1/2} 
\end{align*}
$$

5-31

The CDF due to combined loading, $K_I(g)$, was then evaluated using Equation 5-32.

$$
K_I(g) = K_I^P + gK_J^S 
$$

5-32

Graphs of $K_I$ versus $L_I$ were plotted for axial loading and pressure loading. On each graph, the CDF due to combined loading (based on the $\rho$ and $g$-parameters), primary loading and full elastic-plastic FEA were displayed.
5.5 Numerical modelling results

5.5.1 Option 2 FAC

Figure 5-14 – Option 1 and Option 2 Failure Assessment Curves for the parent material.

Figure 5-14 shows the Option 2 FAC, with an Option 1 FAC for comparison. In general the Option 1 FAC is conservative across most of the $L_r$ range. Between $L_r = 0.7$ and $L_r = 1.1$, the Option 1 FAC is most conservative; in this range the work-hardening capacity of the stainless steel material is accounted for in the Option 2 FAC. Up to $L_r = 1.4$ both FACs are coincident. Beyond $L_r = 1.4$ the Option 2 FAC is less conservative than the Option 1 FAC, but the cut-off at $L_r^{\text{max}} = 1.54$ prevents assessment points from falling in the region where the two FACs begin to diverge significantly.
5.5.2 Axial loading

Figure 5-15 shows the increase of CDF with applied load for the axially-loaded numerical model. The CDF due to the residual stress field is apparent at low $L_r$, where the combined loading models produce secondary CDFs at zero applied load. It can be seen that at low values of $L_r$, the combined loading models show an increase in CDF at a rate similar to primary loading only.

Beyond $L_r = 0.4$, the combined loading model based on the $\rho$-parameter shows an enhancement of CDF with increasing $L_r$, diverging from the other models. The FE result runs parallel to the primary loading model up to $L_r = 0.6$, where the onset of plasticity serves to redistribute the residual stresses, reducing their contribution to the total CDF. This process continues until the CDF equals that due to primary loading at $L_r = 1.25$, where the residual stresses have been overwhelmed by the primary load.

The $g$-parameter model follows the FE model closely; up to $L_r = 0.8$ the $g$-parameter model is very slightly non-conservative, but becomes conservative at higher $L_r$. For strongly strain-hardening materials, a critical condition for crack initiation is most likely to be met at high $L_r$, where the $g$-parameter is conservative. These results suggest that for this geometry, the $g$-parameter approach accounts for redistribution of residual stresses more effectively than the $\rho$-parameter, tending towards the FE result more closely at high $L_r$ than the $\rho$-parameter approach.
The different levels of conservatism of the various approaches is apparent when the weld fracture toughness is taken into consideration. The critical condition of crack initiation, defined as the point at which the CDF equals the measured weld fracture toughness, $K_{j,0.2}$, is dependent on the assessment approach used. The CDF due to primary loading only and the CDF calculated by the FE model are in good agreement at the weld fracture toughness, where $L_r = 1.26$. The $g$-parameter approach shows that the critical CDF is reached at $L_r = 1.23$, which agrees with the FE result to within 2.5%. The $\rho$-parameter approach is more conservative, reaching the critical CDF at $L_r = 1.15$; almost 10% less than the FE and primary loading results.

However, the measured weld fracture toughness is not strictly valid, as described in Section 4. To satisfy the fracture toughness test standards, $J_{\text{max}}$ would be used to define the critical condition of crack initiation. For comparison, $J_{\text{max}}$ for the weld material has been calculated as an equivalent value, $K_{j,\text{max}}$. At $K_{j,\text{max}}$ the FE and primary loading results agree to within 2%, at $L_r = 1.2$. The $g$-parameter result is
similar, matching $K_{j_{max}}$ at $L_r = 1.15$, within 3% of the FE result. However, the $\rho$-parameter result reaches $K_{j_{max}}$ at $L_r = 1.06$, 12% lower than the FE result.

It is worth noting that at the collapse load (where $L_r = 1$), the CDF evaluated using all three combined loading approaches lies below both $K_{j_{max}}$ and the measured weld fracture toughness. The $\rho$-parameter approach was most conservative; at $L_r = 1$, $K_j = 190 \text{ MPa}\sqrt{\text{m}}$ (85% of $K_{j_{max}}$ and 67% of the weld fracture toughness). The $g$-parameter evaluated $K_j = 150 \text{ MPa}\sqrt{\text{m}}$ at $L_r = 1$. The FE parameter evaluated $K_j = 140 \text{ MPa}\sqrt{\text{m}}$ at $L_r = 1$, showing that the FE and $g$-parameter were in agreement to within approximately 7%. The FE result and the $\rho$-parameter differ by 30%, reflecting the high conservatism of the $\rho$-parameter.
5.5.3 Pressure loading

Figure 5-16 shows the change in CDF with an increasing pressure load. It is clear that for pressure loading, the $\rho$-parameter approach is much more conservative than for the axially-loaded cases. The $\rho$-parameter approach predicts the CDF to equal the weld fracture toughness at $L_r = 1.1$, substantially lower than $L_r = 1.3$ for the FE model. Unlike the FE and $g$-parameter approaches, the $\rho$-parameter doesn’t account for residual stress redistribution until the CDF exceeds the fracture toughness of the weld material.

![Graph of Crack Driving Force versus applied primary load, for the pressure loaded model.](image)

As with the axially-loaded models, the $g$-parameter approach is in good agreement with the FE approach up to $L_r = 0.8$, beyond which the $g$-parameter becomes conservative.
The CDF due to primary loading only and the CDF calculated by the FE model agree to within 1.5% at the weld fracture toughness, where \( L_r = 1.30 \). The \( g \)-parameter approach shows that the critical CDF is reached at \( L_r = 1.27 \), which agrees with the FE result to within 2.5%. The \( \rho \)-parameter approach is significantly more conservative, reaching the critical CDF at \( L_r = 1.09 \), approximately 18% less than the FE results.

At \( K_{j_{\text{max}}} \) the FE and primary loading results agree to within 1%, at \( L_r = 1.2 \). The \( g \)-parameter result is similar, equalling \( K_{j_{\text{max}}} \) at \( L_r = 1.15 \), within 4% of the FE result. The \( \rho \)-parameter result reaches \( K_{j_{\text{max}}} \) at \( L_r = 1.03 \), 12% lower than the FE result.

Similarly to the axially-loaded model, at the collapse load, \( L_r = 1 \), the CDF evaluated using all three combined loading approaches lies below both \( K_{j_{\text{max}}} \) and the weld fracture toughness. The \( \rho \)-parameter approach was most conservative; at \( L_r = 1, K_j = 205 \text{ MPa}\sqrt{\text{m}} \) (91% of \( K_{j_{\text{max}}} \) and 72% of the weld fracture toughness). The \( g \)-parameter evaluated \( K_j = 138 \text{ MPa}\sqrt{\text{m}} \) at \( L_r = 1 \). The FE parameter evaluated \( K_j = 109 \text{ MPa}\sqrt{\text{m}} \) at \( L_r = 1 \), showing that the FE and \( g \)-parameter differed by approximately 22%. The FE result and the \( \rho \)-parameter differed by 62%, showing that the \( \rho \)-parameter is considerably over-conservative for the pressure-loaded model.

The axially-loaded models and pressure-loaded models showed some similarities. Where the crack driving force equalled the weld fracture toughness and \( K_{j_{\text{max}}} \), the FE and primary loading results were in closest agreement. The \( g \)-parameter result agreed to within 4% of the FE result for both axial and pressure load models. However, for both loading modes, the \( \rho \)-parameter is significantly more conservative than the other approaches. Under pressure loading, the \( \rho \)-parameter is significantly more conservative than for axial loading. This is at its most apparent when considering the CDF at \( L_r = 1 \), where the \( \rho \)-parameter result for \( K_j \) is almost as high as \( K_{j_{\text{max}}} \). This is due to the \( \rho \)-parameter failing to account for the redistribution of residual stress in the ductile material at relatively low values of \( L_r \). In Figure 5-16 it can be seen that the CDF doesn’t begin to diminish due to residual stress redistribution until \( L_r \) exceeds 1.1. Historically the \( \rho \)-parameter was developed for a range of materials, often with limited ductility; thus for many materials, the \( \rho \)-
parameter approach is suitably conservative. However, for highly ductile materials, the $g$-parameter would appear to be more appropriate, calculating CDF values in good agreement with full-elastic-plastic FEA, yet remaining slightly conservative.

5.6 Summary

Three approaches were used to determine the elastic-plastic CDF for a postulated 10 mm deep, fully circumferential crack in an NG pipe weld, under axial and pressure loading. The results show that for very ductile materials, the $g$-parameter is a much less pessimistic approach to calculate the CDF, in comparison to the $\rho$-parameter. However, it is also apparent that for the pipe weld used in this project, both the fracture parameters of $K_{\text{max}}$ and the weld fracture toughness, $K_{J_{0.2}}$, exceed the CDF at the plastic limit load of the pipe. This suggests that for ductile pipe welds, extensive plastic deformation would occur before significant crack growth takes place, indicating that plastic collapse solutions may be a more suitable failure parameter for such components; this is discussed further in Section 6.
6 Discussion

6.1 Residual stress measurements

6.1.1 Comparison of residual stress measurement techniques

The through-wall residual hoop and axial stress profiles of the NG pipe weld are shown in Figure 6-1 and Figure 6-2 respectively. Each figure shows the residual stress as measured using the original DHD method, the ND method and the modified DHD performed on the SE(B) blanks. It is clear that the ND results are in much better agreement with the modified DHD results, than the original DHD measurements (shown as a green line on Figure 6-1 and Figure 6-2). As discussed in Section 4, the hoop stress measured using ND at 15 mm depth is unlikely to be representative of the true residual hoop stress at that point, but is included in Figure 6-1 to illustrate the difference in ND and DHD results at that point. Excepting the point at 15 mm depth, the residual hoop stresses measured using ND agree with the modified DHD results to within 12%. The residual axial stress ND measurements agree with the modified DHD results to within 16%.

Figure 6-1 - Through-thickness hoop stress measurements using DHD and ND.
The ND and DHD measurements were performed at independent locations on the weldment; subtle variations in measured residual stress may exist due to slight differences in the exact cross-sectional profile of the weld at each measurement location. Overall the ND results and the modified DHD results performed on the blank specimens are broadly consistent with each other.

The original DHD measurements, however, do not correlate well with the ND and modified DHD results. In the case of the axial stress measurements, the original DHD measurements have a similar form to the other measurements; the stress at the OD and ID are similar and the areas of tensile and compressive residual stress align with those of the other measurements. In the case of the hoop residual stress, the stress measured at the OD is similar, but the stress at the ID, and the location of the tensile and compressive stress zones, do not agree. Most importantly, from a structural integrity point of view, the original DHD measurements fail to capture the peak tensile and compressive residual stresses; in the case of the axial residual stresses, the peak stress measured using the original DHD technique disagrees with the stress measured using the modified DHD technique by 65%. Similarly, the
residual hoop stress measured by the original and modified DHD techniques differ by 31%.

An explanation for the discrepancy between the original DHD and the DHD performed on the blank specimens may be found through consideration of plasticity in the weld material. As highlighted in Section 2.3.7, a key assumption upon which DHD is based is that the measured residual stress field does not generally exceed the yield strength of a material. Therefore, residual strains relax elastically upon trepanning, enabling elastic equations to be used to calculate the residual stress. However, for very ductile materials, such as the stainless steel pipe weld, a distinct yield point (and associated yield stress) does not exist. The 0.2% proof strength, often interpreted as a measure of yield strength, is used for straightforward comparison with other materials. In very ductile materials, plastic deformation takes place at stresses well below the 0.2% proof strength of the material. There is no linear elastic stress-strain relationship up to the proof strength; indeed, the 0.2% proof strength is simply an arbitrary measure of the stress required to cause 0.2% plastic strain in a material. The proof strength gives no indication of the plastic strain path of the material prior to reaching that stress level. In the case of ductile materials, assuming that stress relaxation is entirely elastic can therefore lead to misleading estimates of residual stress.

Such matters are further complicated by the plastic history of the material as a result of the welding process. The temperature, speed and path of each weld pass affect the plastic deformation in each preceding pass. As each pass is deposited, a degree of plastic strain is introduced into the material, which then changes on cooling. A subsequent weld pass can re-heat the original pass, tempering the weld, changing the strain levels in the original pass further. Re-melting of the original weld pass can take place, removing some plastic strain through annealing. The level of strain change will depend on the location of the original pass relative to other weld passes, its proximity to a free surface or the bulk of the parent material, all of which affect the thermo-mechanical history of the original weld pass. Welds and the adjacent parent material therefore have a complex strain history; to accurately map the level of elastic and plastic strain in any given region of a weldment is not trivial, if not impossible. Validated numerical models of the welding process may be the only approach to fully mapping the strain history of the material in a welded joint.
By considering plasticity, the limitations of the DHD technique are revealed. In Figure 6-1 and Figure 6-2 the greatest discrepancies between the original DHD measurements and the other measurements are to be found at the highest residual stress levels, where plastic strains are at their highest. This limitation of the DHD technique has been investigated by Mahmoudi et al. (2009), who performed numerical simulations of DHD on aluminium alloys. They showed that for relatively low stresses, up to 150 MPa, DHD can capture the residual stress levels with a good degree of accuracy. However, for higher stress levels, where plastic strains become more dominant, DHD fails to measure the plastic strains. Mahmoudi et al. (2009) cite one example, where remote stress fields of 100 MPa and 330 MPa were applied to aluminium cylinders; DHD failed to distinguish between the two, calculating a stress of 100 MPa in both cylinders.

Mahmoudi et al. (2009) then modelled a development of the original DHD technique, known as incremental Deep Hole Drilling (iDHD). The process is explained below, outlined in Figure 6-3:

a) Reference bushes are attached to the outer and inner surfaces of the component. A 1.5 mm diameter reference hole is gun-drilled component
b) An air probe travels down the reference hole, measuring the diameter of the hole through a number of angles to determine the profile of the hole
c) A cylinder of material surrounding the reference hole is trepanned using EDM, 0.2 mm from the outer surface
d) The air probe is sent down the reference hole again to measure the change in diameter of the reference hole as a result of the relaxation of residual stresses from the trepanning process. The probe measures the entire length of the reference hole every time an increment is trepanned.
e) Steps (c) and (d) are repeated in steps of 0.2 mm until the reference bush is reached on the inner surface
f) The change in diameter of the reference hole is used to calculate the levels of residual stress that were originally present in the component.

The incremental technique enables strains to relax in a more gradual fashion. Although the same elastic relations are used to calculate the stress from the relaxed strain in iDHD as DHD, the gradual plastic strain relaxation upon incrementally
trepunneling a component is adequately approximated by an elastic equation. It is upon large plastic relaxation that an elastic description of strain relaxation fails, as is the case with conventional DHD.

Mahmoudi et al. (2009) demonstrated the improvement in stress measurement of iDHD over DHD through numerical modelling: for aluminium alloy cylinders with an applied remote stress field of 330 MPa, the iDHD technique measured a stress of approximately 310 MPa, as opposed to the DHD measurement of 100 MPa. This is shown in Figure 6-4.
Figure 6-4 - Numerical simulation, comparing the stresses measured using DHD and iDHD (labelled as modified DHD) in aluminium alloy cylinders under an applied remote stress field (Mahmoudi et al. 2009).

The work of Mahmoudi et al. (2009), showing that conventional DHD fails to capture high plastic strains, provides an explanation for the failure of DHD to capture the peak tensile and compressive residual stresses in the NG pipe weld. Their work also explains the better correlation between the ND results and the DHD performed on the blank SE(B) specimens. The DHD performed on the specimens was pseudo-incremental in nature, illustrated by Figure 6-5: a hole was gundrilled and its profile was measured. The SE(B) blanks were then machined around the hole, leading to partial relaxation of the elastic and plastic residual strains. The hole profile was then re-measured. The material surrounding the hole underwent wire-EDM to completely relax the strains, after which the hole profile was measured again. By relaxing the strains in a two-step process, the plastic strain relaxation had been measured more accurately than the original DHD measurement. This explains why the two SE(B) DHD measurements are in better agreement with the ND results, and why the greatest discrepancy between the SE(B) DHD and the conventional DHD is greatest in the regions of highest plastic strain. Performing the technique again, but using more increments, may improve the accuracy of the results, but even a two-step process shows significantly higher accuracy than using the conventional technique alone.
The iDHD technique keeps the trepan diameter constant, but incrementally increases the depth at which trepanning is performed. By contrast, the modified DHD technique used on the SE(B) blank specimens incrementally changes the trepan “diameter”, as illustrated in Figure 6-5. Further research into incremental reduction of trepan diameter, and the accuracy of the measured residual stress field may be worthwhile; this is discussed further in Section 8.

To summarise, it would appear that for components containing low residual stresses, or made of materials with low ductility, DHD is adequate for measuring residual stress. However, for residual stresses present in highly ductile and strain-
hardening materials, or those with complex plastic histories, such as welds, the modified DHD method is a more suitable approach.

6.1.2 Comparison of residual stress measurements with published results

The residual stress measurements performed in this project have been compared with those published in the open literature. Bouchard (2007) published residual stress measurements using DHD on a narrow-gap weld. Liu et al. (2011) and Yang et al. (2011) modelled pipe welds of similar materials to this project. The pipe welds featured in the published literature are summarised in Table 6-1.

<table>
<thead>
<tr>
<th>Weld type</th>
<th>Parent material</th>
<th>Weld material</th>
<th>Thickness (mm)</th>
<th>No. weld passes</th>
<th>Residual stress measurements</th>
</tr>
</thead>
<tbody>
<tr>
<td>This project</td>
<td>NG butt weld</td>
<td>304L</td>
<td>308L</td>
<td>33</td>
<td>-</td>
</tr>
<tr>
<td>Liu et al. (2011)</td>
<td>“</td>
<td>304L</td>
<td>316L</td>
<td>70</td>
<td>73</td>
</tr>
<tr>
<td>Yang et al. (2011)</td>
<td>“</td>
<td>304L</td>
<td>308L</td>
<td>76</td>
<td>42</td>
</tr>
</tbody>
</table>

Table 6-1 – Summary of pipe welds tested or modelled in published work.

Figure 6-6 and Figure 6-7 compare the through-wall residual stress profiles measured using DHD on the blank specimens with those of Bouchard (2007). Because the original weld residual stress fields measured using the hoop and axial blank specimens were similar, the two results were averaged to produce a single through-wall weld residual stress field for straightforward comparison with the data in the literature. The through-thickness stress profiles in the literature were measured from the pipe ID to the OD. The stress profiles measured in this project have been reversed to match the literature, and the distance through thickness normalised for straightforward comparison.
Figure 6-6 compares the through-wall residual hoop stress profile of this project with the NG girth-weld of Bouchard (2007). The proof strength of the weld material used by Bouchard lies within 3% of the proof strength of the weld material used in this project. As a result, it was unnecessary to normalise the plots in Figure 6-6 and Figure 6-7 with respect to the proof strength. The NG pipe weld of Bouchard was 62 mm thick (almost twice that of the one used in this project), but with similar material properties: Bouchard’s weld material was made of 316L stainless steel, with a proof strength of 446 MPa, compared to 308L weld material with a proof strength of 459 MPa for this project. Bouchard’s parent material, 316H steel, had a proof strength of 287 MPa, compared to 304L steel with a yield strength of 284 MPa. The results show good agreement, with similar areas of tensile and compressive residual stress. At the OD, the measurements are in good agreement, although at the ID there is a discrepancy. However, the DHD technique is susceptible to surface effects; the discrepancy between the results may be due to the measurement technique rather than a genuine physical effect.
Figure 6-7 - Through-wall residual axial stress profile comparing the DHD results on the blank SE(B) specimens with Bouchard’s (2007) DHD results.

Figure 6-7 compares the axial residual stress profile of Bouchard (2007) and this project. There is generally good agreement between the two sets of data. The areas of tension and compression are similar, although residual stresses can vary by up to 100 MPa at the surface. The variations between the stress profiles in Figure 6-7 may be attributed to differences in the pipe geometry and weld parameters. However, both sets of results generally support each other as both show residual stresses of similar magnitude in similar locations through the pipe wall.

Figure 6-8 and Figure 6-9 compare the weld residual stress profiles measured across the centreline of the NG welds. Liu et al. (2011) and Yang et al. (2011) simulated the residual stresses at the outer surface. Liu et al. (2011) also performed centre-hole drilling on the outer surface. These results are compared to the cross-weld residual stress profile measured 10 mm below the outer surface in this project, using ND.

Figure 6-8 shows that both the simulated and experimental results have common features; the hoop residual stress reaches a highly tensile level at approximately
20 mm either side of the weld centreline, falling rapidly towards zero at 50 mm either side of the centreline. Liu et al.'s (2011) simulation overestimates the hoop stress at the centreline; Yang et al. (2011) underestimate it. Liu et al.'s (2011) surface measurements show a form similar to those of this project; two local maxima at the weld-parent boundary, with a local minimum on the weld centreline. Liu et al.'s (2011) measured hoop stresses are approximately 60% of the magnitude of those measured in this project; however, at the surface one would expect to see lower residual stresses due to a loss of constraint.

![Figure 6-8](image)

Figure 6-8 – Hoop residual stress profile measured across the weld centreline, comparing the simulated results of Yang et al. (2011) and Liu et al. (2011) with the DHD measurements performed on the blank SE(B) specimens.

Figure 6-9 compares the simulated and measured axial residual stresses across the pipe weld centrelines. The simulations and experimental results all show profiles of a similar form. Once again, the ND results of this project record higher stresses, due to the increased stress triaxiality 10 mm below the OD. All the results show that peak axial residual stresses lay within 20 mm either side of the weld centreline, but unlike the hoop stresses, the axial residual stress does not fall away rapidly, instead reducing to a level between 100 MPa and 200 MPa up to a distance of 100 mm from the centreline.
In general the published data support the residual stress profile measured across the weldment: the location of the peak residual stresses coincide, and the rate at which the residual stress fields diminish away from the weld centreline are consistent.

![Graph showing residual stress profile](image)

**Figure 6-9** - Axial residual stress profile measured across the weld centreline, comparing the simulated results of Yang et al. (2011) and Liu et al. (2011) with the DHD measurements performed on the blank SE(B) specimens.

### 6.1.3 Summary

The technique based on the $\sin^2 \psi$ method was used to measure weld residual stresses using ND; the results were in good agreement with the modified DHD data and with published data. The modified DHD technique would appear to be a more suitable approach for measuring residual stress in ductile components, due to its ability to account for plastic strains better than conventional DHD. Although the iDHD technique has been developed to account for plastic strains more accurately, iDHD is a more complex process requiring more data analysis, yet may not lead to significantly more accurate results than those obtained using the modified DHD.
technique performed on the blank SE(B) specimens. However, simulation of the welding process through numerical modelling may be the only approach capable of capturing the complex strain history of the material in a weld; such an approach may be validated using experimental data obtained through ND or the modified DHD method.

6.2 Fracture toughness measurements

6.2.1 Comparison with published data

The initiation fracture toughness parameter $J_{0.2BL}$ was used to compare the results from this project to those of Ould et al. (2009) and Yang et al. (2011), shown in Table 6-2.

<table>
<thead>
<tr>
<th>Material</th>
<th>Yang et al. (2011)</th>
<th>Ould et al. (2009)</th>
<th>This project</th>
</tr>
</thead>
<tbody>
<tr>
<td>Test temperature (°C)</td>
<td>177</td>
<td>316</td>
<td>300</td>
</tr>
<tr>
<td>Specimen type</td>
<td>C(T)</td>
<td>C(T)</td>
<td>C(T)</td>
</tr>
<tr>
<td>Specimen thickness, B (mm)</td>
<td>25.4</td>
<td>25.4</td>
<td>25.4</td>
</tr>
<tr>
<td>Specimen width, W (mm)</td>
<td>50</td>
<td>50</td>
<td>50</td>
</tr>
<tr>
<td>Average $J_{0.2BL}$ (kJ/m$^2$)</td>
<td>1307</td>
<td>934</td>
<td>~ 400</td>
</tr>
</tbody>
</table>

Table 6-2 - Summary of fracture toughness measurements made by Ould et al. (2009) and Yang et al. (2011) and from this project.

A key difference between the test method used in this project and that of Ould et al. (2009) and Yang et al. (2011) was the temperature at which the tests were conducted. The effect of elevated temperature serves to reduce the measured fracture toughness, although the fracture toughness measured by Yang et al. (2011) at 316°C is more than double that of the toughness measured by Ould et al. (2009) at 300°C. The results from this project, tested at ambient temperature, lie between those of Yang et al. (2011) and Ould et al. (2009).

The relatively high fracture toughness of the weld metal, measured in this project, may be due to the reduced constraint at initiation provided by the 10 mm thick
SE(B) specimens. In contrast, Yang et al. (2011) and Ould et al. (2009) used 25.4 mm thick C(T) specimens, which, having a higher constraint, would lead to a more conservative measure of fracture toughness. This would appear to be true for the results of Ould et al. (2009), recording a lower fracture toughness than that measured in this project. However, the results of Yang et al. (2011), despite using higher constraint specimens at an elevated temperature, record a higher fracture toughness than the specimens tested in this project. Further discussion of the validity limits applicable to these fracture toughness tests follows later in this section.

The test method used by Ould et al. (2009) and Yang et al. (2011) was based upon the unloading compliance technique, where a single test specimen is loaded, unloaded and reloaded to cause incremental crack growth. In contrast, this project employed the multi-specimen technique, where a fixed amount of crack growth took place in each test specimen. The unloading compliance technique is sensitive to set-up and preparation. For example, should loading pins suffer from a lack of lubrication, or bind on poorly-finished surfaces, the loads required to extend the crack may be higher than for a well-lubricated test pin with clean surfaces. Specimens require bedding-in before crack extension is performed; different test-houses use different bedding-in cycles, which can affect the load at which crack initiation takes place. Interpretation of the unloading lines on a load-displacement trace, to estimate the crack growth with each reload, is dependent on the operator. These factors can cause some variation in the $J$-$R$ curves obtained using similar test specimens made of similar materials. This can make $J$-$R$ curves hard to replicate, and can explain the differences between the $J$-$R$ curves shown in Figure 6-10, comparing the results of this project with those of Ould et al. (2009) and the bounding curves of Yang et al. (2011). The weld material tested in this project showed a higher toughness up to a crack extension of approximately 0.5 mm, before showing similar levels of fracture toughness to the specimens tested by Yang et al. (2011). Although the specimens, test methods and test temperatures vary significantly between the different groups, it can be seen that the $J$-$R$ curves of the three sets of results are of a similar order of magnitude.
6.2.2 Validity of fracture toughness tests

Figure 6-11 shows the $J$-resistance curve obtained for the hoop and axial weld specimen fracture toughness tests. In terms of compliance with test standards and procedures, the $J$-R curve satisfies the limit for crack growth, $\Delta a_{\text{max}}$, but exceeds the crack driving force limit, $J_{\text{max}}$. The purpose of the limits is to ensure high constraint levels are maintained at the crack tip; if these limits are exceeded, constraint loss may have taken place, leading to a non-conservative fracture toughness measurement.
Figure 6-11 - J-resistance curve for the weld material, combining the data from the hoop and axially-orientated specimens.

In the case of the weld specimens, $J_{\text{max}} = 232 \text{ kJ/m}^2$. All of the data points lie above this value, and would be censored if the test procedures were strictly followed. The validity of a fracture toughness test can only be obtained once testing is complete. It is therefore difficult to design test specimens of an appropriate size to satisfy the validity limits, unless an estimated fracture toughness of the material can be obtained prior to testing; for example, by using data from earlier tests on similar specimens. For all of the above data points to be valid, a $J_{\text{max}}$ equal to 886 kJ/m$^2$ is required. To provide the level of constraint required to satisfy this limit, SE(B) specimen dimensions of thickness $B = 34$ mm and width $W = 67$ mm would be required. For a pipe with a wall thickness of 33 mm, such a specimen could not be machined. This problem was also encountered by Ould et al. (2009) and Yang et al. (2011): despite using thicker 25.4 mm (1 inch) C(T) specimens, both groups exceeded the $\Delta a_{\text{max}}$ and $J_{\text{max}}$ limits. Testing thicker specimens would normally be undertaken to raise the $J_{\text{max}}$ value to produce valid fracture toughness measurements. However, Ould et al. (2009) also tested C(T) specimens of 50.8 mm thickness (larger than the 33 mm thickness pipe welds used in this project); these results also exceeded the $J_{\text{max}}$ limit. The $J$-$R$ curve of Ould et al. (2009), shown in
Figure 6-10, was fitted to data obtained from both 25.4 mm thick and 50.8 mm thick C(T) specimens.

To overcome the problem of using an appropriately sized specimen, a modified specimen could be designed by welding extra material onto the pipe weld so that a larger specimen may be extracted. However, the effects of joining additional material by welding may introduce further complexity into the material microstructure, potentially influencing the toughness measurement. An alternative method may be to extract a C-ring specimen from the pipe, introducing a starter crack through EDM. Such an approach would have to be supported through the use of numerical modelling prior to testing, to ensure that an appropriate level of constraint was maintained during a test to obtain a conservative measure of fracture toughness.

6.2.3 Choice of curve-fit

The use of curve-fitting techniques can influence the derived values of fracture toughness for a material. For the fracture toughness tests performed on the parent and weld materials, Table 4-6 lists the different fracture toughness parameters based on the type of curve-fit applied to the test data.

<table>
<thead>
<tr>
<th>Specimen material/orientation</th>
<th>Offset power-law curve fit</th>
<th>Power-law curve fit</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>$J_{0.2}$ (kJ/m²)</td>
<td>$J_{0.2BL}$ (kJ/m²)</td>
</tr>
<tr>
<td>Parent, hoop</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Weld, hoop</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td>Weld, axial</td>
<td>387</td>
<td>531</td>
</tr>
<tr>
<td>Weld, combined</td>
<td>395</td>
<td>521</td>
</tr>
</tbody>
</table>

Table 6-3 - Summary of fracture toughness parameters using two curve fits.

The choice of curve-fit is set by test standards and procedures, although different interpretations of the standards can lead to different results. For example, both BS7448 (2005a) and ESIS P2-92 (1992) recommend an offset power law curve fit, shown in Equation 6-1.
Where \( m, l \) and \( x \) are curve fit constants. Both BS7448 and ESIS P2-92 permit \( m = 0 \), so that if an offset power law curve cannot be fitted, a power law curve fit may be used instead. For the data obtained in this project, the difference in measured fracture toughness can vary by approximately 6% depending on the curve-fit used. For consistency, the power law curve fit was used as it was possible to obtain \( J-R \) curves for all of the test specimens using this approach.

### 6.2.4 Influence of specimen orientation on measured fracture toughness

The crack driving force required to cause 0.2 mm of tearing, \( J_{0.2} \), in the axial and hoop weld specimens agree to within 4% of each other. Performing a power law curve fit to both the axial and hoop data yielded a value of \( J_{0.2} = 371 \, \text{kJ/m}^2 \), which lies halfway between the \( J_{0.2} \) values calculated separately. The weld material microstructure, analysed in the crack planes for both hoop and axial specimens, were the same, with little variation in grain size, shape and distribution. The presence of retained residual stress in the axial specimens, which is discussed further in Section 6.3, would appear to have had little effect on the measured fracture toughness. The axial specimens showed a slightly higher measure of \( J_{0.2} \) than the hoop specimens. Had the residual stress had an effect on measured toughness, the axial specimens would have exhibited lower fracture toughness than the hoop specimens. For this narrow-gap weld, the retained residual stress, and thus specimen orientation, has had no observable effect on the measured fracture toughness.

### 6.2.5 Summary

The \( J \)-resistance curve of the weld material tested in this project is similar to published results, despite different test methods and conditions being used. Obtaining results which satisfy the validity limits set out by test standards and procedures is not always possible when testing ductile materials, despite designing specimens as large as can be accommodated by the original component. The development of novel test specimens may be required to obtain a more
conservative measure of fracture toughness in narrow-gap welds: the narrow-gap geometry prevents extraction of weld test specimens that satisfy current standards and procedures. The choice of curve-fitting technique used to obtain a $J$-R curve can influence the measured fracture toughness; for consistency, the same method should be applied across all data if possible. Testing SE(B) specimens aligned with the hoop and axial directions of the pipe weld has showed that in an austenitic stainless steel narrow-gap weld, the orientation of the test specimen would appear to have had no effect on the measured fracture toughness.

6.3 Retained residual stress in SE(B) specimens

6.3.1 Residual stress relaxation

The modified DHD measurements showed that a significant level of residual stress was retained in the axial SE(B) specimen. However, in the hoop SE(B) specimen, little residual stress was retained. This effect may be explained by comparing these results with the observations of Altenkirch et al. (2009), who measured residual stress in sectioned welded plates.

Altenkirch et al. (2009) referred to the longitudinal and transverse components of weld residual stress. For the pipe weld, the longitudinal component corresponds to the hoop stress, and the transverse component to the axial stress. Altenkirch et al. (2009) observed that both longitudinal and transverse sectioning of a component to produce test specimens caused the longitudinal residual stress distribution to shift downwards to balance the stress distribution across a specimen. However, in the case of transverse residual stresses, neither longitudinal nor transverse sectioning caused a significant change in transverse residual stress, due to the transverse residual stresses self-equilibrating over a relatively short distance across the weld centreline. However, Altenkirch et al. (2009) observed that both transverse and longitudinal residual stresses would relax if a specimen was substantially smaller than the component from which it was extracted.

The SE(B) specimens extracted from the pipe weld were much smaller than the original component: the length and thickness of each specimen were less than 10%
of the original length and circumference of the pipe. It is therefore reasonable to expect that both the longitudinal and transverse components of residual stress would change on extraction of the specimens.

### 6.3.2 Extraction of axial SE(B) specimen

Figure 6-12 shows how the axial SE(B) specimen was extracted from the pipe weld, using a combination of transverse and longitudinal sectioning. For this specimen, the dominant sectioning effect took place in the longitudinal direction, indicated by the large red arrows, to extract a specimen 10 mm thick.

![Figure 6-12 - Extraction of axial SE(B) specimen from the pipe weld.](image)

The DHD measurements show that the axial SE(B) specimen retained a significant level of residual stress upon extraction. This is because the longest dimension of the specimen, the length, provided sufficient material to surround the weld, serving to preserve most of the transverse residual stress field. This agrees with the observations of Altenkirch et al. (2009); the transverse residual stresses were generally insensitive to sectioning, due to the transverse residual stresses self-
equilibrating over a relatively short distance across the weld centreline. It is worth noting that the DHD measurements were performed only in the centre of the SE(B) specimen; as shown in Figure 6-9, the axial residual stress is estimated to decay substantially within 50 mm of the weld centreline, which may also be the case in the extracted SE(B) specimen.

In contrast, the hoop stresses were relaxed upon extraction of the axial SE(B) specimen. The thickness of the specimen in the hoop direction was 10 mm, providing insufficient material to constrain the weldment to prevent residual stress relaxation from taking place.

6.3.3 Extraction of hoop SE(B) specimen

Figure 6-13 shows how the hoop SE(B) specimen was extracted from the pipe weld, through a combination of transverse and longitudinal sectioning. For this specimen, the dominant sectioning effect took place in the transverse direction, indicated by the large blue arrows, to extract a specimen 10 mm thick.

Figure 6-13 - Extraction of hoop SE(B) specimen from the pipe weld.
In contrast to the axial SE(B) specimen, the hoop SE(B) specimen did not retain a significant level of axial or hoop residual stress. Altenkirch et al. (2009) observed that axial stresses are largely insensitive to sectioning; however, if the extracted piece was very small compared to the body of the component, stress relaxation would still take place. In the hoop SE(B) specimen, the thickness of the specimen in the axial direction, 10 mm, is substantially smaller than the original length of the pipe. Extracting the specimen, which was made entirely of weld material, left no surrounding parent material to constrain the residual stress, causing relaxation of the axial (transverse) stresses.

Despite the largest dimension of the hoop SE(B) specimen – the length – being aligned with the hoop direction of the pipe, very little hoop residual stress was retained. Altenkirch et al. (2009) observed that the hoop (longitudinal) stresses are sensitive to both transverse and longitudinal sectioning. Firstly, the length of the extracted specimen, 92 mm, is substantially smaller than the original length of the pipe weld; the original circumference of the pipe was over 1100 mm in length. Secondly, the large reduction in transverse width (from 200 mm to 10 mm), caused further hoop stress relaxation, due to a reduction in constraint in the axial direction.

6.3.4 Estimating retained longitudinal residual stress

To describe the relaxation effect quantitatively, the two empirical relations derived by Altenkirch et al. (2009) were applied to the extracted SE(B) specimens. The two equations, introduced in Section 2, are repeated here. Equation 2-54 relates the retained longitudinal stress, \( \sigma_{\text{long}} \), to the original residual stress field before the weldment was sectioned longitudinally.

\[
\sigma_{\text{long}} = \sigma_0 \left[ 1 - \exp \left( \frac{1}{3} \left( - \frac{l}{w_t} \right) \right) \right] \tag{6-2}
\]

Where \( w_t \) is the width of the tensile residual stress peak across the weldment, \( \sigma_0 \) is the original residual stress (strictly the residual stress in a theoretically infinitely long weld) and \( l \), the length of the extracted specimen.
Equation 2-55 relates the retained longitudinal stress upon transverse sectioning to the original weld residual stress. The width of the extracted specimen is denoted by the parameter $w$.

$$\sigma_{\text{long}} = \sigma_0 \left[1 - \frac{w_t}{w}\right]$$

Because extracting the SE(B) specimens involved both longitudinal and transverse sectioning, Equation 2-54 and Equation 2-55 were applied simultaneously to the original residual stress field data. The ratio of retained longitudinal residual stress to original longitudinal residual stress, $\sigma_{\text{long}}/\sigma_0$, was calculated.

For Equation 2-54 and 2-55, Altenkirch et al. (2009) defined the half-peak width, $\frac{1}{2}w$, as the width from the weld centreline to the distance where the residual stress fell to zero. The ND measurements across the pipe weld centreline did not extend to an area where the residual stresses fell to zero. Extrapolating the ND data showed that the residual stress would diminish at approximately 100 mm from the weld centreline, equalling a tensile peak width of 200 mm. Setting $w_t = 200$ in Equation 2-54 and Equation 2-55 gave erroneous results.

The hoop stress at the weld centreline, 400 MPa, halved in value at approximately 37 mm from the centreline. This corresponds to a stress peak width, $w$, equal to 75 mm, illustrated in Figure 6-14. Applying this value to Equation 2-54 and Equation 2-55 produced stress relaxation values in good agreement with the experimental results, which are shown in Table 6-4.
Table 6-4 shows that the relaxation equation must be chosen with care. In the case of the hoop specimen, the measured longitudinal stress is in good agreement with Equation 2-54 for longitudinal sectioning. However, Equation 2-55, does not hold for transverse sectioning of the hoop specimen.

In the case of the axial specimen, the measured retained hoop stress agrees with Equation 2-55, for transverse sectioning, but does not agree with the longitudinal sectioning relation described by Equation 2-54.

The stress relaxation equations derived by Altenkirch et al. (2009) are empirical, derived from experiments quite different to those of this project. It is worth noting
that the above equations were derived from thin plates sectioned in one direction only. The residual stress across the weld was measured at the mid-thickness of each plate. In contrast, this project involved extracting specimens through a combination of longitudinal and transverse sectioning in a thick pipe weld. The residual stress across the pipe weld was measured at 10 mm depth from the OD, approximately one-third through the thickness of the pipe.

However, the relaxation equations have estimated the longitudinal residual stress relaxation in close agreement with experimental results, when an appropriate relaxation equation was used. In the case of the hoop specimen, a greater level of transverse sectioning was performed than longitudinal sectioning to produce the test-piece. Equation 2-54, describing longitudinal stress relaxation, agrees with the experimental results for the hoop specimen.

For the axial specimen, a greater level of longitudinal sectioning was performed than transverse sectioning to create the test-piece. Equation 2-55, estimating transverse stress relaxation, agrees with the experimental results for the axial specimen.

It may be concluded that the retained residual stress equations developed by Altenkirch et al. (2009) can be used to estimate the level of longitudinal residual stress retained in a test specimen, with the following caveats:

- The tensile peak width, \( w_t \), of the residual stress field must be interpreted with care,
- The choice of relaxation equation depends on the dominant sectioning mode, for example, if transverse sectioning was dominant in producing a test-piece, the longitudinal retained stress equation should be used (and vice-versa).

Further work is required to test the equations developed by Altenkirch et al. (2009). For example, measuring the residual stress halfway through the thickness of the pipe weld to establish the tensile peak width at mid-thickness would allow for a more direct comparison of results. Extracting test pieces from a pipe weld using one sectioning mode only would also provide a more effective form of comparison with the results of Altenkirch et al. (2009).
The above approach applies to longitudinal (hoop) stresses only. To calculate the retained transverse (axial) stresses in a test specimen, the approach outlined in the R6 procedure, applied by Hurlston et al. (2012), is discussed below.

### 6.3.5 Estimating retained transverse residual stress

The transverse (axial) stress relaxation of the weld residual stress field upon extraction of material from the pipe weld was calculated using equations from the R6 procedure, outlined in Section 2, as used by Hurlston et al. (2012). The original weld residual stress field, measured through DHD performed on the axial and hoop SE(B) specimens, was separated into three components: self-equilibrating, bending and membrane stress. The results for the axial SE(B) specimen are shown in Figure 6-15.

![Calculated retained axial residual stress in the axial SE(B) specimen.](image-url)
Figure 6-15 shows the axial residual stress profile through the thickness of the pipe wall, from the OD to the ID. The 6 mm and 26 mm bounds mark the area in which the main bulk of the extracted SE(B) specimen lay. Between these bounds the retained residual stress measured in the specimen is in agreement with the self-equilibrating component of stress calculated using the R6 approach. The calculated and measured peak tensile and compressive residual stresses agree to within 15% of each other within the bounded region, diverging outside this region. The bending component of stress intersects the measured retained stress at the membrane stress and the 26 mm bound, showing good agreement between experiment and calculation.

The retained transverse (axial) stress, measured in the hoop SE(B) specimen, is compared to the R6 approach, Figure 6-16. The calculated self-equilibrating component of stress is not in agreement with the retained residual stress measured in the specimen; the retained residual stress does not show contrasting areas of highly tensile and highly compressive residual stress. However, within the 6 mm to 26 mm bounds, the retained stress in the specimen is of a similar magnitude to the calculated membrane component of stress, approximately 40 MPa.

The discrepancy between the experimental values and the calculation may be due to the loss of axial constraint in the hoop SE(B) specimen. The reduction in width from the original pipe length, 200 mm, to the specimen thickness, 10 mm, has resulted in a large loss of constraint, which is not accounted for in the R6 approach. However, this is consistent with the observations of Altenkirch et al. (2009): retained stresses can only be reliably calculated when sectioning is moderate. For specimens much smaller than the original component, the large loss of constraint cannot be accounted for by the approaches used in R6 or by Altenkirch et al. (2009). However, in the case of the hoop SE(B) specimen, the membrane stress may be used to estimate the small amount of transverse stress retained in a specimen which has experienced a large loss of constraint.
6.3.6 Retained residual stress crack driving force

The most significant amount of residual stress retained in a test specimen was the axial residual stress present in the axial SE(B) specimen. In such a specimen, the retained residual stress may contribute significantly to the CDF in a fracture toughness test, influencing a toughness measurement.

To estimate the crack driving force due to the retained residual stress, the axial SE(B) specimen was treated as an edge-cracked plate of finite width. The crack depth, $a$, was chosen to be half the specimen width, 10 mm, within the pre-crack growth limits defined by fracture toughness test standards. The SINTAP handbook of Stress Intensity Factors and Limit Load Solutions (Al Laham 1999) was used to calculate the CDF, using Equation 6-4.

$$K = YZA\sqrt{a}\left(\sigma_0 + \frac{F}{W}\right)$$

Figure 6-16 - Calculated retained axial stress in the hoop SE(B) specimen.
Where \( a \) is the crack depth, \( \sigma_0 \) is the uncracked body stress at the mouth of the crack and \( W \) is the specimen width. \( F \) is defined in Equation 6-5.

\[
F = \int_0^a \frac{2(W - x)^2}{\pi W} \left[ a \cos \left( \frac{x(W - a)}{a(W - x)} \right) \right] \frac{d\sigma}{dx} \, dx \tag{6-5}
\]

Where \( x \) is the distance along the crack depth from 0 to \( a \). The term \( YZA \) is defined in Equation 6-6.

\[
YZA = \frac{\sqrt{\pi} \left( 1 + 2\left( \frac{a}{W} \right) \right) U}{\left( 1 - \left( \frac{a}{W} \right) \right)^{3/2}} \tag{6-6}
\]

Where \( U \) is defined in Equation 6-7.

\[
U = 1.12078 - 3.68220 \left( \frac{a}{W} \right) + 11.9543 \left( \frac{a}{W} \right)^2 - 25.8521 \left( \frac{a}{W} \right)^3 \\
+ 33.09762 \left( \frac{a}{W} \right)^4 - 22.4422 \left( \frac{a}{W} \right)^5 + 6.17836 \left( \frac{a}{W} \right)^6 \tag{6-7}
\]

Equations 6-4 to 6-7 were evaluated numerically, yielding a stress intensity factor of \( K = 78 \) MPa√m. The measured fracture toughness of the axial weld specimens was 290 MPa√m; the CDF due to the retained residual stress in the axial specimen equalled 27% of the measured fracture toughness of the weld material. However, the measured fracture toughness of the hoop weld specimens was 282 MPa√m. The hoop specimens have been shown to retain little of the original weld residual stress field, yet show similar fracture toughness to the axial test specimens. This suggests that the retained residual stress in the axial specimens has had little effect on the measured fracture toughness of the weld material. The ductility of the material has enabled large plastic strains to redistribute the residual stress in the specimen, reducing the residual stress contribution to the total crack driving force. For less ductile materials which exhibit brittle failure characteristics, the retained residual stress in a test specimen could, however, have a significant effect on the measured fracture toughness of the material, causing failure at lower loads than would be the case without retained residual stress present. This could lead to a measured
fracture toughness which is over-conservative for the material, influencing the outcome of structural integrity assessments.

6.3.7 Summary

The level of residual stress retained in a test specimen can be estimated using a combination of methods. The empirical equations derived by Altenkirch et al. (2009) can be used to estimate the longitudinal residual stress retained in a test specimen, provided that the stress peak width is interpreted correctly. In the case of longitudinally-orientated specimens, the longitudinal relaxation equation holds, and for transversely-orientated specimens, the transverse relaxation equation holds. The R6 approach adopted by Hurlston et al. (2012) can be used to estimate the transverse retained residual stresses in a test specimen. For transversely-orientated specimens, the self-equilibrating component of stress agrees with experiment, and for longitudinally-orientated specimens, the membrane component of stress may be used to estimate the retained transverse stress.

A significant amount of transverse residual stress can remain in a transversely-orientated test specimen; the retained residual stress can contribute to approximately one-third of the CDF required to initiate fracture. However, in the narrow-gap weld material, the retained residual stress had little effect on measured fracture toughness, but could have a significant effect on materials which exhibit brittle failure characteristics. For ductile materials, testing axially-orientated specimens may be the most appropriate approach, for several reasons. Firstly, axial specimens can be extracted using all of the wall-thickness of the pipe, maximising the specimen width and therefore maintaining constraint levels as high as possible, increasing the conservatism of a fracture toughness test. Secondly, many more axially-orientated specimens can be extracted from a pipe weld than hoop-orientated specimens, which is a more economical use of material and provides more test specimens for multi-specimen testing, leading to more data points for improved J-resistance curve fits.
6.4 Numerical modelling

The results in Section 5.5 showed that for both axial and pressure loading of a fully-circumferential crack in the NG pipe weld, the $\rho$-parameter is the most conservative interaction parameter. The $g$-parameter was shown to be effective at reducing the conservatism of the $\rho$-parameter, whilst remaining slightly more conservative than the elastic-plastic FE result. However, the choice of interaction parameter may not be as significant as the choice of failure parameter. For example, the fracture toughness of the weld material, $K_{\text{f}0.2}$, may be chosen as the critical condition, or $K_{\text{fmax}}$ if valid fracture toughness data cannot be obtained. Alternatively, the collapse load of the component, $L_\epsilon = 1$, may be chosen as the critical condition for the purpose of structural integrity assessment.

In the case of the fully-circumferentially cracked pipe model, the CDF equal to both $K_{\text{f}0.2}$ and $K_{\text{fmax}}$ exceeded the CDF at $L_\epsilon = 1$, regardless of the interaction parameter used. This suggests that using a collapse load solution may be a more appropriate approach to assess the structural integrity of components similar to the NG pipe weld. A limitation of this investigation was the modelling of only one type of defect in the pipe weld. The fully-circumferential external crack modelled in this project can occur in practice, for example, if a circumferential weld pass does not successfully fuse with the neighbouring material. However, there are many other types of defect that can exist in a pipe weld. For example, partially circumferential cracks can develop at the ID or the OD. Axial cracks may initiate, with hoop stresses making the most significant contribution to the CDF. Further work, modelling a range of defects in the pipe weld under service loads may serve to establish whether a collapse load solution is the most suitable failure parameter for NG pipe welds. Should further work show that for certain defects, the CDF exceeds $K_{\text{f}0.2}$ at loads lower than the collapse load, then the $g$-parameter would appear to be the most suitable choice of interaction parameter for assessing the structural integrity of NG pipe welds.
7 Conclusions

This chapter summarises the main conclusions from this research project, which investigated the weld residual stress state and tensile and fracture properties of a primary circuit narrow-gap girth welded austenitic stainless steel pipe and their influence on structural integrity.

1. The residual stress field was characterised in three dimensions using deep hole drilling and neutron diffraction. The maximum hoop and axial components of residual stress were of similar magnitude to the weld material yield stress. The high tensile hoop residual stress region was located between 2 and 20 mm from the outer surface of the pipe weld, and extended up to 30 mm either side of the weld centreline. This tensile stress was balanced by a region of compressive stress close to the inner surface of the pipe.

2. A novel neutron diffraction scanning technique, similar to the $\sin^2 \psi$ method used in X-ray diffraction, was applied to measure the hoop and radial residual strains, and hence residual stresses, in the pipe weld. As a result, no window had to be cut into the pipe wall to allow the neutron beam to access the measurement volume, thus ensuring that the original weld residual stress field was not affected by the measurement technique.

3. A modified deep hole drilling technique was applied to blank fracture mechanics specimens machined from the weld. The residual stress field was measured more accurately than through the conventional deep hole drilling technique, without resorting to the more complex and time-consuming incremental deep hole drilling method.

4. The tensile properties of the parent and weld material were characterised; the 0.2% proof strength was 284 MPa and 459 MPa respectively, showing that the narrow-gap weld was approximately 60% overmatched. Both the parent and weld materials showed similar work-hardening behaviour, with a work hardening exponent of $n = 2.2$ and 2.8 respectively.
5. The fracture toughness properties of the parent and weld materials were characterised. The parent material exhibited an initiation fracture toughness value at 0.2 mm of tearing, $K_{f0.2} = 308 \text{ MPa}\sqrt{\text{m}}$. For the weld material, both the hoop and axially-orientated test specimens had the same measured fracture toughness of $K_{f0.2} = 284 \text{ MPa}\sqrt{\text{m}}$, showing that the initiation toughness is independent of orientation in the pipe weld. However, as concluded below, the validity of these fracture toughness measurements is an issue due to the limited size of fracture mechanics specimen that could be extracted from the pipe weld.

6. Extracting a test specimen from a narrow-gap weld large enough to comply with the validity limits set out by test standards and procedures was not possible. The maximum fracture toughness for the weld specimens permitted by the ESIS P2-92 procedure was $K_{f\text{max}} = 207 \text{ MPa}\sqrt{\text{m}}$, 31% lower than the measured value of $K_{f0.2} = 284 \text{ MPa}\sqrt{\text{m}}$. This difference is likely to relate to the loss of crack-tip constraint as a result of using small fracture mechanics specimens.

7. A significant amount of residual stress can be retained in an extracted test specimen; in the case of the axially-orientated weld specimens, the peak retained residual stress had a magnitude of 54% of the weld material 0.2% proof stress. However, the retained residual stress had no observable effect on the measured fracture toughness. This is likely to result from large plastic strains redistributing the residual stresses in the uncracked ligament prior to crack initiation. In a less ductile material, this level of retained residual stress may unduly influence the measurement of fracture toughness.

8. The longitudinal and transverse components of residual stress retained in a test specimen can be estimated, provided that significant constraint loss has not taken place upon extraction of the specimen. A comparison of estimated and measured retained residual stress using the Altenkirch approach showed agreement within 3%; for the R6 stress partitioning approach, the measured and estimated retained stresses were found to be within 15% of each other.
9. For the structural integrity assessment of components containing defects under combined primary and secondary loading, using the $g$-parameter to account for the influence of plasticity on the secondary crack driving force reduced the level of conservatism of the current $\rho$-parameter approach used in the R6 procedure. The $g$-parameter enabled the calculation of a combined crack driving force in close agreement with that predicted by a full elastic-plastic finite element analysis of a circumferential defect in the narrow-gap pipe weld.

10. The assessment of the failure condition in the narrow-gap pipe weld containing a circumferential defect, 10 mm in depth, showed that a plastic collapse solution may be a more appropriate failure parameter than assessing the critical crack driving force against the fracture toughness at initiation of ductile tearing; such an approach may lead to more straightforward structural integrity assessments of similar components.
8 Further work

8.1 Residual stress measurement

Performing only one neutron diffraction linescan across the pipe weld limited information of the residual stress state through the pipe wall. Further characterisation of the residual stress field, by performing linescans through the wall thickness, would characterise how the peak residual stress changes in the zone close to the weld centreline. A measurement of the weld residual stress field at the mid-thickness of the pipe wall would provide a more direct comparison with the residual stress field measured at the mid-thickness of welded plates by Altenkirch et al. (2009), which is discussed further in Section 8.2. To perform linescans further into the wall thickness would require an access window for the neutron beam to be machined from the pipe weld, potentially relaxing some of the original weld residual stresses. However, the linescan at 10 mm depth could be repeated, and compared with the original measurements performed in this project using the $\sin^2 \psi$ technique, to establish whether any relaxation had taken place. An access window would allow for the measurement of residual strains to be taken directly, negating the need for the $\sin^2 \psi$ approach, making data acquisition times faster and requiring less data analysis following the experiments.

Further investigation of the modified DHD method would be of interest. Extracting the SE(B) blanks from the pipe weld was analogous to trepanning the material around the reference hole with a large trepan diameter, permitting some strain relaxation before full relaxation of residual stress took place upon EDM of the specimen. Trepanning in two stages yielded residual stress measurements in good agreement with the ND results; further work, investigating the number of trepanning steps, and the effect of different trepan diameters, could be undertaken to establish the sensitivity of the technique.

A possible method is illustrated in Figure 8-1, where the trepan diameter is reduced three times to relax the residual strains incrementally. Such an approach would involve removing a large volume of material from a component, which would turn the DHD method into a destructive measurement technique. However, the modified technique may prove suitable for measuring residual stress in thick walled
components, where the use of alternative techniques, such as ND, is not possible. For research purposes, the technique could be used to measure the residual stress in decommissioned components, or weld mock-ups, similar in configuration to in-service components.

8.2 Retained residual stress in test specimens

The relations derived by Altenkirch et al. (2009) were applied to the blank SE(B) specimens extracted from the pipe weld, to estimate the residual stress retained in the specimens. However, extracting the specimens involved both longitudinal and transverse sectioning, whereas Altenkirch et al. (2009) derived their relations by performing one sectioning mode at a time. Further research could involve sectioning a pipe weld, one sectioning mode at a time, to measure the retained residual stress, and comparing the results with the relations derived by Altenkirch.
et al. (2009), to see if they apply to pipe welds (as opposed to thin plates). Measuring the residual stress in sectioned pieces could be performed either by ND or DHD; signal attenuation issues in ND would not be an issue as the path length through a small specimen is short. Performing such tests would also provide an opportunity to test the stress peak width, \( w_t \), as a characterising parameter of the residual stress field, or, for pipe welds, whether a different characterising parameter should be used. Further characterisation of the original weld residual stress field through the thickness of the pipe wall, as discussed in Section 8.1, could be used to investigate the variation in the measured stress peak width, and to explore the effect of a change in stress peak width on the prediction of retained residual stress in extracted specimens.

### 8.3 Fracture toughness tests

As discussed in Section 6, obtaining valid fracture toughness measurements of the weld metal required fracture mechanics specimens of a thickness significantly larger than the wall thickness of the pipe weld to be machined. To overcome this problem, the approach taken in the ADIMEW project (Taylor et al. 2006) may be used, where weld metal is deposited on the OD of the pipe, building up layers of material from which larger fracture toughness specimens can be machined, as shown in Figure 8-2.

![Weld build-up technique](image)

Figure 8-2 - Weld build-up technique used to accommodate full-size fracture mechanics specimens from a pipe weld (Taylor et al. 2006).
However, the welding process may alter the original weld microstructure to the point where it is no longer representative of its original condition. Alternatively, additional material could be joined to the OD using a technique such as Electron-Beam welding, which has less impact on the neighbouring material microstructure, due to the very low heat input, resulting in less distortion and weld residual stress across the joint (Lacki & Adamus 2011). Thus, only the properties of the material in a small region local to the electron-beam weldment would be affected by the welding process, which should not affect the results of a fracture mechanics test performed on the original weld material.

Alternative approaches to accurately derive fracture toughness measurements from small volumes of material could be taken. For example, a modelling approach could be developed to explore size effects on the fracture toughness of thick sections of ductile material, by testing specimens ranging in size, investigating how ductile fracture models, such as the Gurson damage mechanics model (Gurson 1977), predict the results.

The ductile material assessed in this project showed no observable difference in the measured fracture toughness between specimens of different orientation. Performing a set of tests comparing specimen orientation and measured fracture toughness on a less ductile material, where there is less plasticity to redistribute residual stresses, may help to establish whether retained residual stress and specimen orientation can influence the measured fracture toughness in less ductile materials.

8.4 Numerical modelling

The 2-D numerical model of the pipe weld is limited in scope for assessing the structural integrity of a narrow-gap pipe weld containing a range of postulated defects. Only a fully-circumferential defect was modelled, yet many other types of defect can exist in such components. Performing a full 3-D numerical model of the pipe weld would enable a more realistic defect to be modelled – such as a partially circumferential surface crack, or an axial crack, which could be used to further validate the $g$-parameter as a suitable interaction parameter for combined primary
and secondary loading. Furthermore, the collapse load of the pipe weld containing a range of defects could be investigated, to determine if the collapse load is a more suitable failure parameter than the measured fracture toughness of the material, which may lead to more straightforward structural integrity assessments of narrow-gap welds fabricated from ductile materials.

Equally, simulating the NG welding process, and validating against residual stress measurements, may help to develop an understanding of the plastic history of the weld and parent materials upon welding, and aid in the development of more sophisticated models of the weld residual stress field and its effects on the behaviour of defects under combined loading.
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Appendix A  Test records
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<th>Specimen type</th>
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Table A-1 – Test matrix
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Table A-2 – List of specimens extracted from received material
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<th>UTS (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of area (%)</th>
<th>Proof/UTS ratio</th>
<th>0.2% proof strength (MPa)</th>
<th>UTS (MPa)</th>
<th>Flow strength (MPa)</th>
<th>Elongation (%)</th>
<th>Reduction of area (%)</th>
<th>Proof/UTS ratio</th>
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<td>592</td>
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Table A-3 – Tensile test data
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Table A-4 – Initial crack length measurements in fracture toughness specimens
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Table A-5 – Crack extension measurements in fracture toughness specimens
Table A-6 – Fracture toughness test data calculated from end-point measurements

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<th>Specimen code</th>
<th>Test temp. (°C)</th>
<th>Thickness Gross, B_g (mm)</th>
<th>Thickness Net, B_n (mm)</th>
<th>Width, W (mm)</th>
<th>Initial crack length, a_0 (mm)</th>
<th>Crack extension, Δa (mm)</th>
<th>Measured b_0 (mm)</th>
<th>a_0/W</th>
<th>Area under CMOD/Load curve, U</th>
<th>J (kJ/m²)</th>
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<td>10.04</td>
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Table A-7 – Summary of fracture toughness data analysis results
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Table A-8 – Summary of fracture toughness parameters using different curve fits
Appendix B  Cutting plans
Figure B-1 - Cutting plan for parent tensile specimens and parent hoop fracture toughness specimens
Figure B-2 – Cutting plan for weld tensile specimens and weld hoop fracture toughness specimens
Figure B-3 - Cutting plan for weld tensile specimens and weld axial fracture toughness specimens
Figure B-4 – Cutting plan for diffraction comb
Appendix C  Specimen drawings
Figure C-1 - Round tensile specimen drawing
Figure C-2 – SE(B) specimen drawing
Figure C.3 – Side-grooved SE(B) specimen drawing
Figure C-4 – Axial SE(B) blank specimen drawing
Figure C-5 – Hoop SE(B) blank specimen drawing
Figure C-6 – Diffraction comb drawing
Appendix D  Specimen and test photographs
Figure D-1 – SE(B) specimen prior to fatigue pre-cracking

Figure D-2 - Detail view of notch and integral knife-edges

Figure D-3 - SE(B) specimen after etching to confirm location of notch in weld material
Figure D-4 - Detail view of notch on weld centreline

Figure D-5 - SE(B) specimen after fatigue pre-cracking, with side-groove

Figure D-6 – Side-grooves, showing included angle and 20% reduction in specimen width
Figure D-7 – Fracture surface of a parent material SE(B) specimen after testing

Figure D-8 – Fracture surface of a weld material SE(B) specimen after testing
Figure D-9 - SE(B) specimen undergoing fracture toughness testing

Figure D-10 – SE(B) specimen undergoing post-test fatigue cracking